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INFLUENCE OF CHEMICAL COMPOSITION ON RUPTURE TEST PROPERTIES AT 1500° F OF FORGED CHROMIUM-COBALT-NICKEL-IRON BASE ALLOYS

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SUMMARY

The influence of systematic variations of chemical composition on rupture properties at 1500° F was determined for 62 modifications of a basic alloy containing 0.15 percent carbon, 1.7 percent manganese, 0.5 percent silicon, 20 percent chromium, 20 percent nickel, 20 percent cobalt, 3 percent molybdenum, 2 percent tungsten, 1 percent columbium, 0.12 percent nitrogen, and the balance iron. These modifications included individual variations of each of 10 elements present and simultaneous variations of molybdenum, tungsten, and columbium. All specimens were forged to bar stock, solution-treated at 2200° F for 1 hour, and aged at 1400° F for 24 hours using procedures to minimize effects of prior history. The rupture strengths for 100 hours and at least extrapolated values to 1000 hours were obtained. The results of a similar study at 1200° F were previously reported in NACA Report 1058.

All of the elements can be varied individually between quite wide limits without significantly changing the rupture properties. This feature of the results could be partially responsible for the difficulty of correlating chemical composition with properties at high temperatures from data in the literature. There was no evidence to indicate that composition variation within usual commercial limits contributes materially to scatter bands in properties.

Carbon could be increased to about 0.40 percent with little change in strength although elongation in the rupture test would be greatly reduced. Increased silicon would be detrimental to strength. The amount of carbon and manganese must be above minimum values to avoid low strength and low elongation, respectively. Alloys with 10 percent of nickel or cobalt have properties similar to those of the basic alloy. Reduction of chromium, molybdenum, tungsten, or columbium below the values in the basic alloy would be detrimental to strength. Further increases are of no benefit and in the case of columbium are detrimental. Alloys containing about 8 percent of molybdenum plus tungsten without the presence of columbium would be as strong as the basic alloy. Alloys

without columbium, however, would have low elongations after fracture in 100 hours while all columbium-bearing alloys would have higher elongations except in the presence of increased carbon.

From a fundamental viewpoint, little could be deduced regarding the mechanism by which composition affects rupture test properties. Creep resistance had a greater effect on rupture strength than elongation to fracture, although there was a definite increase in rupture strength from increased elongation at constant creep strength. The data suggest that a complex interrelation exists between the effects of carbon, molybdenum, tungsten, and columbium and creep resistance. The high elongation of the alloys containing columbium appeared to be due to its grain-refining power and suppression of precipitation in grain boundaries.

The results were similar to those previously obtained at 1200° F, except that a suppression of strengthening from molybdenum and tungsten by columbium at 1500° F was not so evident at the lower temperature. The continued increase in rupture strength observed at 1200° F by increasing chromium to 30 percent was not evident at 1500° F.

There are a number of limitations to the results. Particularly important was the lack of consideration of the effects of compositional variations on fabricating characteristics and the other general properties which are as essential as strength at high temperatures.

INTRODUCTION

The exact influence of chemical composition on the properties of heat-resistant alloys at high temperatures is not clear from published data. Consistent correlations of properties to composition cannot be made. For this reason some uncertainty exists regarding restrictions on composition limits in commercial practice and the fundamental mechanisms by which composition influences properties. For these reasons an investigation was initiated for the purpose of providing systematic data for compositional variations in forged alloys stemming from a nominal composition of 20 percent chromium, 20 percent nickel, 20 percent cobalt, 3 percent molybdenum, 2 percent tungsten, and 1 percent columbium. Stress-rupture tests at 1200° and 1500° F were used to evaluate strength at high temperatures. A previous report (reference 1) presented the results from the tests at 1200° F and the present report extends the data to 1500° F.

The procedures used were selected to provide results as independent as possible of other effects known to influence properties of such alloys at high temperatures and to clarify, or at least to provide the background

necessary for clarifying, the fundamental mechanisms by which composition controls properties. The investigation was intended to establish general principles which could be extended to all alloys rather than to describe specifically the particular alloys used for experimental purposes.

The properties of alloys, of the type under consideration, at high temperatures are not a function of composition alone. Whenever such alloys are used or tested at high temperatures, the results are also a function of the conditions of preparation since melting practice, methods of working the metal to finished form, and final heat treatments appear to be interrelated to compositional effects. Thus, even when compositional effects and final heat treatments are carefully controlled, the results are almost invariably still subject to the prior-history effects. The situation is further complicated by the relative properties depending on the temperature, stress, total deformation, and time period involved in evaluation of properties.

The alloys tested were prepared using procedures intended to minimize prior-history effects. Care was used to keep melting and hot-working conditions constant. The final heat treatment was a solution treatment at 2200° F for 1 hour followed by aging at 1400° F for 24 hours. This treatment was known to minimize prior-history effects for the base alloy. Other treatments were not used because the testing program required for a systematic investigation involving prior-history effects would be prohibitively large.

In undertaking the investigation it was recognized that there was little chance of developing new compositions having properties outstandingly better than those available in commercial alloys. The thorough covering of compositional effects on an empirical basis by both industry and governmental agencies in this country and abroad leaves little chance of such developments. The investigation was intended rather to determine the contribution of composition variation to the substantial scatter bands in properties of such alloys; to indicate the properties obtainable from "leaner" alloys which may be desirable from strategic-alloy or cost considerations; and to determine the efficiency of the various alloying conditions as a possible guide for specifications.

The work was conducted by the Engineering Research Institute of the University of Michigan under the sponsorship and with the financial assistance of the National Advisory Committee for Aeronautics as part of an investigation of the fundamental metallurgy of heat-resisting alloys used in propulsion systems for aircraft.

PROCEDURE

The basic alloy composition studied was as follows:

Carbon, C, percent	0.15
Nitrogen, N, percent	0.12
Manganese, Mn, percent	1.7
Silicon, Si, percent	0.5
Chromium, Cr, percent	20
Nickel, Ni, percent	20
Cobalt, Co, percent	20
Molybdenum, Mo, percent	3
Tungsten, W, percent	2
Columbium, Cb, percent	1
Iron, Fe, percent	Balance

The influence of composition was evaluated on 62 modifications of this alloy. The variations of each element individually were:

C, percent	0.08, 0.40, 0.60
N, percent	0.004, 0.08, 0.18
Mn, percent	0, 0.30, 0.50, 1.0, 2.5
Si, percent	1.2, 1.6
Cr, percent	10, 30
Ni, percent	0, 10, 30
Co, percent	0, 10, 32
Mo, percent	0, 1, 2, 5, 7
W, percent	0, 1, 5, 7
Cb, percent	0, 2, 4, 6

In addition, Mo, W, and Cb were varied simultaneously in steps of 2 percent from 0 to 4 percent. The amount of Fe compensated for variation in total alloy content in all of the alloy modifications.

Each alloy was analyzed chemically for the element varied. In addition, complete analyses were made for each heat until satisfactory evidence of reproducible composition was obtained. Thereafter the composition was checked only occasionally.

Considerable care was used to keep the alloy preparation uniform for each alloy as is described in detail in reference 1. In brief the alloys were melted in an induction furnace as 9-pound heats using Armco iron and commercial ferroalloys as melting stock and calcium silicide as a final deoxidizer. The ingots were hammer forged between 2200° and 1800° F to 0.45-inch rounds using swaging dies for reductions from 1-inch rounds on down. Prior to machining the stock into rupture specimens, all bars were solution-treated for 1 hour at 2200° F, water-quenched,

and then aged for 24 hours at 1400° F in order to minimize the effects of prior history.

Rupture tests at 1500° F were made on specimens 0.250 inch in diameter with a 1-inch gage length machined from the heat-treated bar stock. The tests were conducted in individual stationary units with the load applied by a simple beam acting through a system of knife edges. Usually three or four tests at various stresses were made on each alloy and were of sufficient duration to establish the 100-hour rupture strengths and to estimate the 1000-hour rupture strengths. Time-elongation data were taken during the rupture tests by the drop-of-the-beam method and in the case of many of the longer time tests by means of modified Martens type extensometers with a sensitivity of 0.00005 inch per inch.

Metallographic examinations were made on all of the longest-time rupture specimens after testing at 1500° F. All metallographic specimens were electrolytically etched in 10 percent chromic acid.

Even though care was exercised to maintain all phases of alloy preparation and testing constant, it was recognized that some variation from heat to heat was inevitable. Rupture tests were made on four separate heats of the basic alloy to indicate the reproducibility. The range of values thus obtained was used to indicate significant variations in properties due to changes in chemical composition.

RESULTS

The actual compositions of the alloys obtained by chemical analysis are shown in table I. In general, the actual analyses agreed as closely as could be expected with the intended compositions. In those cases where actual analyses were not made, table I was left blank, although the blanks signify that the composition was intended to be that of the basic alloy. Check analyses from time to time indicated this assumption to be valid.

The reproducibility in rupture properties of the four heats of the basic alloy is summarized by the data for stress against rupture time in figure 1. These results indicate that the stress for rupture in 100 hours would have to vary more than 2500 psi in order to be significant of a real effect from composition. The corresponding value for rupture in 1000 hours was about 1500 psi. Figure 2 shows the range in minimum creep rates obtained during rupture testing of the four basic heats. This figure thus indicates the range which must be exceeded before the effect of composition on creep resistance under rupture test conditions can be considered significant.

The results of the rupture tests are given in table II. In addition to the usual data from rupture tests this table includes the minimum creep rates measured from the time-elongation curves of the rupture tests. The stresses for rupture in 100 and 1000 hours indicated by curves of stress against rupture time are shown in table II, although the actual curves are not included in the report. The stresses for rupture in 1000 hours were considered as extrapolated when the longest-duration test was less than 500 hours. Double-logarithmic curves of stress and minimum creep rate were plotted to obtain the stresses for creep rates of 0.01 and 0.1 percent per hour given in table II. These two values together with the creep rate at 13,000 psi were included to determine if composition was affecting rupture strength by increasing creep resistance under rupture testing conditions. These creep "strengths" together with the total elongation to fracture are considered to be a measure of the two mechanisms by which composition changed rupture strength.

The influence on rupture test properties of the systematic variations in composition for the individual elements is shown by figures 3 to 12. The results from simultaneous variations of Mo, W, and Cb are shown by figures 13 to 21. Summaries of the results are incorporated in figures 22 to 24. The significant features of these figures are outlined in the following sections. In the analysis of the results, only those variations outside the ranges for the basic alloy were considered significant, although in many cases the uniform trend in properties with changes in composition suggests that more significance than that is warranted for the data.

Carbon

Rupture strength was increased by C up to 0.40 percent (see fig. 3). Further increase resulted in a significant decrease in strength. Low C was detrimental to and increased C beneficial to the 1000-hour rupture strength.

Additions of C increased rupture strength by increasing creep resistance. This was sufficient to offset a pronounced decrease in total elongation to fracture with increasing C content.

Nitrogen

Alloys varying in N content from 0.004 to 0.18 percent all had rupture strengths within the band for the basic alloy (see fig. 4). The elongation at fracture of the two extremes, however, was significantly higher. It appears that for N contents less than about 0.05 percent low creep resistance was partially offset by increased elongation to fracture which kept the rupture strength up.

Manganese

Figure 5 shows that Mn variations from 0 to 2.5 percent had little significant effect on rupture properties. When the Mn was below 0.5 percent the elongation to fracture tended to be low and the creep resistance high.

Silicon

Increasing the Si from 0.5 to 1.6 percent progressively and significantly lowered rupture strength (see fig. 6). The lowering of rupture strength was due primarily to decreased creep resistance.

Chromium

A 10-percent-Cr alloy had significantly lower rupture strength than that of the basic alloy as the result of lowered creep resistance (see fig. 7). There did not appear to be a significant difference in rupture strength between the basic alloy and the 30-percent-Cr modification.

Nickel

The data for Ni variations from 0 to 30 percent (see fig. 8) indicated a maximum strength for about 10 percent. The Ni-free alloy had definitely inferior strength while the 10-percent-Ni alloy was very close to being significantly better than the basic alloy. Both creep resistance and elongation to fracture varied in the same manner as rupture strength.

Cobalt

There was no difference between the 10- and 20-percent-Co alloys (see fig. 9). The Co-free and the 32-percent-Co alloys, however, had definitely inferior strengths.

Molybdenum

Rupture strength and creep resistance increased slightly with Mo additions up to 3 percent (see fig. 10). Alloys with less than 3 percent of Mo appeared to have slightly, although significantly, lower strength.

Tungsten

Additions of W had effects similar to those of Mo (see fig. 11). Alloys with less than the 2 percent of W of the basic alloy were definitely inferior. Further additions of W did not produce a significant increase in strength.

Columbium

The Cb-free alloy had significantly lower strength and ductility than the basic alloy (see fig. 12). Increasing Cb above the 1 percent of the basic alloy reduced rupture strength somewhat and creep resistance quite markedly. The reduction of creep resistance was offset by higher elongation in the high-Cb heats, however, so that the rupture strengths tended to stay near those of the basic alloy.

Simultaneous Variations in Molybdenum, Tungsten, and Columbium

The influence on the 100-hour rupture strength of varying Mo with constant combinations of W and Cb is shown by figure 13. The increase in strength was marked and progressive for the Cb-free combinations. The W-free alloys were also improved, but to a lesser extent. Combinations of W and Cb were benefited by the addition of 2 percent Mo and in some cases by 4 percent although the degree of improvement was less than for the previously mentioned combinations. It is significant that additions of up to 4 percent Mo were never detrimental. None of the alloys had better strength than the basic 3Mo-2W-1Cb alloy.

The same data are rearranged in figure 14 to show W as the variable. The general pattern of the influence of W was similar to that of Mo except that the amount of increase for a given addition was less.

The influence of Cb, however, was markedly different from either Mo or W as is shown by figure 15. Additions of 2 percent Cb markedly increased rupture strength but 4 percent was detrimental. The alloys with 1-percent-Cb additions suggest that 1 percent would be sufficient to obtain the maximum benefit and that perhaps some of the other alloys would have had higher strengths than those shown with less than 2 percent Cb.

It will be noted that, of all the Cb-free modifications tested, only the 4Mo-4W alloy had strengths equal to those of the basic 3Mo-2W-1Cb alloy.

The accumulative effect of alloy additions on rupture strength is shown by figures 16, 17, and 18. Little further increase results from

Mo or W additions when the total alloy content exceeds 6 to 8 percent. Additions of Cb of 1 or 2 percent raise the strength to a pronounced extent for alloys having less than 8 percent Mo or W. Further additions of Cb are detrimental regardless of the total alloy content.

The data previously discussed have been rearranged in figure 19 to provide a simultaneous comparison of both rupture strength and elongation at 100 hours. There are three sets of curves at each of three alloy levels: each element varied alone, each element varied in the presence of 2 percent of the other two elements, and in the presence of 4 percent of the two elements. It shows:

(1) The strength of alloys with Mo, W, or Cb added alone was inferior

(2) The Cb-free alloys had low elongation or conversely the elongation increased pronouncedly when Cb was added

(3) Additions of Mo in the presence of Cb and W resulted in a further increase in elongation while W in the presence of Mo and Cb slightly reduced elongation

(4) All of the alloys containing 2 or 4 percent Cb, except the 0Mo-0W-2Cb and 0Mo-2W-2Cb modifications, had more elongation at fracture than the basic alloy

The influence of variations of Mo, W, and Cb on the stress for rupture in 1000 hours (see fig. 20) was, in general, similar to that for rupture in 100 hours. There was some change in relative strengths for the higher alloy modifications and more alloys compared favorably with the basic alloy.

The creep resistance of the alloys with simultaneous additions of Mo, W, and Cb (see fig. 21) indicated that Mo and W additions increased rupture strength by increasing creep resistance. Additions of Cb on the other hand apparently did not increase creep resistance except in those alloys in which Mo and W were absent. This latter observation is not strictly correct because reference to table II shows that the addition of 2 percent of Cb increased the creep resistance of alloys containing 2 or 4 percent of Mo or W alone. As figure 21 shows, however, 2 or 4 percent of Cb was detrimental to the 2Mo-2W and 4Mo-4W alloys. This appears somewhat at variance with the results of changing Cb in the basic alloy where a substantial increase in creep resistance resulted from addition of 1 percent of Cb to the 3Mo-2W combination. In this alloy, however, 2 percent of Cb was detrimental to creep resistance. It is therefore suggested that a 1-percent addition of Cb would show a substantial improvement in both creep resistance and rupture strength for the alloys involved in the simultaneous-variation series.

Summarized Effects of Composition on Rupture Test Properties

The effects of individual variations of the elements in the basic alloy on rupture strength and elongation are summarized in figure 22. The stresses for rupture in 100 and 1000 hours resulting from simultaneous variation of Mo, W, and Cb are summarized by figures 23 and 24, respectively. Elongation was omitted in the latter summary inasmuch as those alloys without Cb had low elongation and all containing Cb had high elongation.

The data for the individual variations were presented in groups according to the percentages present in the alloys. The effects of simultaneous variations were summarized by combining Mo and W as the independent variable because the two elements had similar and additive effects. Average curves were drawn through the data for the various Cb contents studied. This procedure may not be correct in detail in that the effect of Mo and W may be different. The trends, however, are shown by the curves and probably represent the reproducibility of the properties from heat to heat.

MICROSTRUCTURAL EFFECTS OF COMPOSITION VARIATION

The original microstructures were examined in detail for reference 1 to help explain the effects of composition on rupture properties at 1200° F. In this investigation the examination was extended to include the structures of the specimens from the most-prolonged tests at 1500° F.

When tested at 1200° F there was little change in structure in most alloys from that produced by the original solution and aging treatment. After testing at 1500° F, however, the general trend was for the precipitate particles in both the grain boundaries and matrix to be larger. The fractures tended to be more intergranular than at 1200° F in those alloys which did not have brittle-type fractures. These general effects are shown by the structures of the basic alloy in figure 25.

The following discussions briefly summarize the structural effects of the various compositional changes as related to the rupture properties.

Carbon

Increasing C increased the amount of insoluble constituent present after solution-treating and the amount of random precipitate which formed on aging. The grain sizes of alloys up to 0.40 percent C were about A.S.T.M number 5 while the 0.60-percent-C alloy had a grain size

of 8. In addition, there were rather numerous large particles of a new constituent which formed on aging the 0.60-percent-C alloy. The structures of the lower-C heats were not substantially altered by testing at 1500° F (see fig. 26). After testing only one type of large particles was present in the 0.60-percent-C material which was distinguished by its ease of etching in comparison with that of the original constituents.

Nitrogen, Manganese, and Silicon

There were no apparent microstructural effects from the variations in N, Mn, or Si either before or after testing at 1500° F.

Chromium

The main difference between the basic alloy (20 percent Cr) and the 10-percent-Cr alloy was the much smaller amount of precipitates in the 10-percent-Cr material. There were, however, substantial differences between the 20- and 30-percent-Cr alloys (see figs. 25 and 27) after rupture testing at 1500° F. As heat-treated, the 30-percent-Cr alloy exhibited a large amount of fine matrix precipitate arranged on crystallographic planes. After testing at 1500° F a large amount of new phase had formed which exhibited all the metallographic characteristics of one of the forms of sigma phase.

Nickel

Alloys containing 10 and 30 percent Ni were similar to the basic alloy after heat treatment and after testing (see fig. 28). The Ni-free alloy, however, contained a considerable amount of delta ferrite as solution-treated which transformed to sigma phase on aging. During testing at 1500° F the grains of sigma phase broke up into angular sigma particles quite randomly distributed. A few small angular particles of an easily etched constituent believed to be sigma phase formed in the 10-percent-Ni alloy.

Cobalt

As heat-treated, the Co-free alloy had a good deal more general precipitation than the basic alloy. There was little difference between the basic alloy and the 10- or 32-percent alloys, except that segregated growth of extremely large grains had occurred in the 32-percent-Co alloy.

During testing at 1200° F a considerable amount of additional precipitation occurred in the Co-free alloy. The other alloys were not changed.

At 1500° F, however, either the precipitation did not occur in the Co-free alloy or the particles agglomerated into a few large particles (see fig. 29). In addition, there were a few small angular particles of an easily etched constituent believed to be a form of sigma phase. A few of these were also present in the 10-percent-Co alloy.

Molybdenum

Varying Mo from 0 to 7 percent in the heat-treated conditions did not change the structure appreciably from that of the basic alloy. After testing at 1500° F the amount of precipitate present increased with Mo content as is shown by figure 30.

Tungsten

The effect of W on the microstructures was similar to that of Mo. After testing at 1500° F all of the W series were similar to the basic alloy except the 7-percent alloy (see fig. 31). In the latter alloy the precipitates were not so agglomerated and more grain boundary precipitate was present.

Columbium

The omission of Cb from the basic alloy resulted in the grain size increasing from 5 to 1 by the A.S.T.M. rating. In addition, the insoluble constituents were practically eliminated and the aging precipitates accumulated on preferred planes within the grains. Increasing the Cb to 6 percent resulted in an increase of insoluble constituent and a further reduction in grain size to number 8 for the 4- and 6-percent-Cb alloys.

After testing at 1500° F the precipitates in the Cb-free alloy remained preferred, in contrast with the high-Cr alloy previously discussed. The precipitated particles were considerably larger than before testing (see fig. 32). The 6-percent-Cb alloy, however, underwent little change in microstructure except for the development of a considerable number of small easily etched "sigma" particles.

Simultaneous Variation of Molybdenum, Tungsten, and Columbium

The Cb-free alloys had comparatively large grain sizes and a slight amount of general but oriented precipitation in the heat-treated condition. Alloys carrying up to a total of 4 percent of Mo plus W had easily etched grain boundaries because of the formation of a considerable amount of

a fine, easily etched constituent during aging. Alloys having more than 4 percent of Mo plus W did not etch so readily at the grain boundaries.

Additions of Cb resulted in the formation of increasing amounts of a globular constituent which could not be dissolved during solution treatment. The grain size was refined and the preferred orientation of aging precipitates was eliminated, the structures being similar to that of the basic alloy.

Increasing amounts of Mo or W in the presence of Cb increased the amount of general aging precipitate.

After testing at 1500° F, as is shown by figures 33 and 34, a large amount of general precipitation occurred in the Cb-free alloys. The precipitate present in the grain boundaries of the alloys with less than 4 percent of Mo plus W etched very easily and were similar to the easily etched phase in the 0.60-percent-C alloy after testing. In those with more Mo plus W the grain boundary constituents etched similarly to those in the matrix.

Those alloys containing Cb underwent some additional precipitation and agglomeration of the precipitated particles during testing at 1500° F. The 4Mo-4W-4Cb alloy developed small particles of the easily etched sigma phase. The Cb tended greatly to reduce the presence of grain boundary precipitates.

Summary of Microstructural Studies

It is difficult to arrive at conclusive reasons from microstructural studies for the observed effects of alloy additions on the rupture properties. Apparently the improved ductility of the Cb-bearing alloys was due to the grain refinement and elimination of grain boundary precipitation by this addition. The presence of delta ferrite in the N-free alloy was accompanied by reduced strength as is the usual behavior when this occurs. Otherwise the evidence seems contradictory; for example:

(1) The increase in C, Cr, Mo, or W was accompanied by increased precipitation. In each case there was an increase in rupture strength due to increased creep resistance, at least up to a limit. A further increase in rupture and creep strength, however, resulted with the addition of Cb to Cb-free alloys even though the amount of precipitation was reduced.

(2) The almost complete alteration of the microstructure of the 30-percent-Cr modification during rupture testing at 1500° F, due to

the extensive formation of sigma phase throughout the matrix, did not significantly change the rupture or creep properties of this alloy modification.

(3) The formation during testing at 1500° F of relatively small amounts of a precipitate resembling sigma phase did not appear to alter the rupture and creep properties of the alloy modifications in which this type of precipitate formed.

The microstructural studies of these alloy modifications indicate that more information is needed relating microstructures to high-temperature properties before a good correlation can be made between microstructure and rupture and creep properties.

DISCUSSION

The data need to be considered from the viewpoints of their limitations, ranges in composition before significant changes in properties result, the mechanisms by which composition probably affects properties, and their relation to the results at 1200° F of reference 1.

Limitations of Results

The main limitation of the data is the restriction to one solution-treated and aged condition. The probability that heat treatment should be adjusted to each individual analysis for either optimum properties or freedom from prior-history effects was not considered. Present experience with the variables affecting rupture properties, however, suggests that the heat-treatment limitation is not too severe. There may have been some variation in solution effect with composition as well as in the amount of precipitation during the aging treatment, but it is doubtful if these effects would be large. It is very probable that changing prior treatments uniformly would tend to show the effects to be uniformly superimposed on the properties measured in this investigation.

Further limitations of the data involve the ranges of compositions studied and the criteria of strength used to evaluate the high-temperature strength. Compositional effects shown are actually limited in that they do not involve simultaneous variations of all of the elements. The tests were limited to rather short-time rupture tests although the indicated relative effects of composition are quite likely to change if the stress level, total deformation, and time periods used as the criteria of strength were changed.

Significance of Data

The results suggest that most of the elements could be allowed to vary over quite wide ranges without affecting the rupture strength significantly. This could be partially responsible for the difficulty of correlating chemical composition with properties at high temperatures from data in the literature. It is important, however, to recognize that the data do not indicate that all of the elements could simultaneously vary widely. This would require work to determine the influence of all of the elements being on the low and high sides simultaneously. The indication of possible variation over wide limits without sacrifice in properties is important, however, from strategic-alloy and cost considerations. Any such changes, however, would have to take into consideration the effects of the compositional changes on fabricating characteristics and other properties which are the over-all requirements for the successful use of an alloy. The results of this investigation, however, do suggest that the following changes might be useful in meeting specific problems:

(1) The C content could be allowed to vary over a considerable range although higher C would undoubtedly increase fabrication problems and would be accompanied by lowered elongation in the rupture test.

(2) A Ni content of less than 20 percent seems quite possible and might improve strength somewhat.

(3) The Co could be reduced with little apparent loss in strength although this trend should probably be checked for the influence of Co variation on response to strain-hardening before acceptance. There is considerable evidence which suggests that Co is beneficial to the improvement and retention of strength by strain-hardening. This is important to the properties obtained with many commercial treatments.

(4) Even better properties might result by decreasing the Cb content to less than 1 percent.

The comments in regard to the individual variation of the other elements involve a difficult viewpoint as evidenced by the following:

(1) The elements N and Mn have little effect on rupture strength. Their control to close limits other than to avoid very small amounts apparently depends on other properties than strength at high temperatures.

(2) The Cr content needs to be close to a minimum of 20 percent.

(3) The 3Mo-2W-1Cb combination of the basic alloy seems to be near the optimum combination. A decrease in either Mo or W would significantly weaken the alloy, but, on the other hand, no great advantage resulted from further increasing Mo or W while increasing Cb was actually detrimental.

Simultaneous variations of Mo, W, and Cb showed that equal rupture strengths to that of the basic alloy could be obtained without the use of Cb by increasing the Mo plus W content to about 8 percent. The Cb-free alloys, however, had comparatively low elongations to fracture and comparatively poor forgeability (see reference 1). The simultaneous variations bore out the indication that little would be gained by increasing Mo, W, or Cb above the 3:2:1 ratio of the basic alloy.

The rather insensitive response of rupture strength to compositional changes indicates it is unlikely that normal variation of composition within usual commercial limits contributes significantly to scatter of data. The spread in properties for the basic heats in comparison with the compositional effects supports this view. The scatter bands observed in commercial production would therefore seem due to prior-history effects rather than compositional variation.

In analyzing the data it has been assumed that values had to be outside of the bands for the basic alloy to be significant although there is no way of knowing whether the individual values were at the middle or the extremes of the band for that composition. The generally consistent variation in properties with composition suggests that more reliance can be placed on the data than the band for the basic alloy suggests. Nevertheless, the reproducibility of properties for alloy modifications should be checked before too much reliance is placed on the data for any particular composition.

Fundamental Effects of Composition

From the data of this investigation very little can be deduced regarding fundamental mechanisms by which composition changes properties. For instance:

(1) Rupture strength is increased by Mo and W in proportion to the amount added in the absence of Cb particularly when considered on an atomic percent basis as in figure 35. This suggests that Mo and W increase strength by solid-solution strengthening (they had little effect on elongation). In the presence of Cb, however, their effect was greatly reduced and saturation was soon attained beyond which no further improvement was obtained. If solution of Mo or W was responsible for their increasing strength it seems difficult to understand their reduced effect in the presence of Cb.

(2) The data indicate that no further increase in strength was obtained by increasing Cb beyond 1 percent. In view of the certain restricted solubility of Cb the improvement obtained from Cb seems rather large. This suggests some factor other than solid solubility is responsible for the effect of Cb. The effect of Cb could be accounted

for on the basis of a combination of increased creep resistance from a limited solubility and the increase in elongation. This, however, does not explain the restriction of the strengthening effect from Mo and W in the presence of Cb.

(3) A complex interrelation between C (and N), Mo, W, and Cb is suggested by the data. This has some support from the increased creep resistance accompanying increased C in the basic alloy. The stabilization of C (and N) by Cb perhaps restricts the strengthening from solid solution of Mo and W in the same manner as restriction of C in the basic alloy reduced strength. There seems to be little to support a saturation effect being rapidly attained in the presence of Cb because of its stabilization of C which, in turn, improved the efficiency of Mo and W additions. A further possibility exists that Mo, W, and Cb influence some other strengthening reaction such as might be due to an effect of Cr.

(4) Additions of Mo and W mainly increased rupture strength by increasing creep resistance. Additions of 1 or 2 percent of Cb increased both creep resistance and elongation to fracture and therefore increased rupture strength by both mechanisms. The 100-hour rupture strengths are plotted against a measure of the creep resistance in figure 36 with elongation to fracture as a parameter. The increase in rupture strength was directly proportional to the increase in creep resistance. The correlation of rupture strength with elongation is not so good although there is an unmistakable trend for increased rupture strength with increased elongation. Creep strength had a much greater effect than elongation.

(5) The mechanism by which Cb so pronouncedly increased elongation to fracture was also uncertain. The refinement in grain size may have been a contributing factor, although it is doubtful that it is the real reason. It appears as if Cb acts by prevention of the development of brittle grain boundaries. A possible mechanism would involve its stabilization effect on C and N combined with a change in precipitation habits in the alloys due either to its presence in solution or to the insoluble precipitates accompanying its presence. The decrease in elongation accompanying increased C contents in the basic alloy supports the interrelation of Cb and C as being the responsible mechanism.

(6) The role of aging during testing at 1500° F is not evident from this investigation. The aging data in reference 2 indicate that unquestionably there was considerable precipitation during testing of the type which results in a decrease in lattice parameter, an effect due to the removal of odd-size atoms from solution. Presumably these odd-size atoms would be largely Mo, W, and C with the added possibility of Cb if there was an appreciable amount in solution. Reference 2 also demonstrated that aging was accompanied by a decrease in creep resistance

at 1200° F. A similar effect of aging has been found at 1500° F from data as yet unreported. It would seem, therefore, that aging during testing should have contributed to a decrease in rupture time during the tests.

There were several features of the microstructural changes during testing which should be considered in the fundamental mechanisms of compositional effects:

(1) The highest-C heat fell off in strength. This was accompanied by a major modification of the carbides during testing. The possibility exists that this was due to reversion during testing to leaner-C-content carbides. This presumably would reduce the Mo and W in solution and thereby reduce creep resistance.

(2) There was some slight evidence of a similar behavior in alloys containing more than 4 percent of W or Mo, although a similar effect on strength was not evident. Changes in composition of the carbides during heat treatment and testing, however, may be a more significant factor than has yet been evident from the work to date.

(3) There were no structural reasons evident to explain the detrimental effect of Si.

(4) The alloy containing 30 percent Cr underwent a pronounced change in microstructure with practically no effect on rupture properties. The microstructure after testing at 1500° F indicated the presence of substantial amounts of sigma phase which is usually found to be quite weak at elevated temperatures.

(5) The delta ferrite originally present in the Ni-free alloy also showed evidence of reversion to sigma phase during testing. In this case a relatively small amount of sigma phase was accompanied by a significantly lower strength. A similar behavior by the Co-free alloy without the presence of sigma phase, however, suggests the possibility that some other factor than the delta ferrite was responsible for the low strength of the Ni-free alloy, and that this factor operated in both alloys.

(6) There were other major effects on rupture properties from compositional changes which could not be accounted for on the basis of microstructures.

Relative Effects of Composition at 1200° and 1500° F

Comparison of the general effects of composition on rupture properties at 1500° F with those reported in reference 1 leads to the following:

(1) The general effect of C was the same at both temperatures except that at 1500° F increasing C was more detrimental to elongation and 0.60 percent of C caused a reduction in strength at 1500° F.

(2) The relative effects of N, Mn, Si, and Ni were the same at both temperatures. Very low N, Mn, and Ni had lower relative strengths at 1500° F than at 1200° F. At both temperatures Si was detrimental to strength but did not increase elongation to fracture at 1500° F as it did at 1200° F.

(3) At 1200° F there was the same increase in strength from 20 to 30 percent of Cr as from 10 to 20 percent. The improvement between 20 and 30 percent was not observed at 1500° F, however, and apparently the beneficial effect of precipitation hardening from the highest Cr content was not retained at 1500° F.

(4) Whereas the data at 1500° F tended to show equal and maximum properties at both 10 and 20 percent of Co, no benefit from Co was observed at 1200° F until 20 percent was added.

(5) Additions of Mo had the same general effect at both temperatures except that smaller amounts than the basic 3 percent were not so detrimental at 1500° F.

(6) At both temperatures W had the same general effects.

(7) The general effects of Cb were the same at both temperatures except that increasing the amount beyond 1 percent was not detrimental to creep strength at 1200° F and did not increase the elongation so much as at 1500° F.

(8) The effects of simultaneous variations of Mo, W, and Cb were very similar at both 1200° and 1500° F. The main differences were that less improvement resulted at 1500° F from the additions of Mo and W and a tendency to reach a saturation effect was noted at about 6 percent of Mo plus W at 1500° F. At 1200° F continued improvement resulted from further additions of Mo plus W so that a number of alloys had rupture strengths superior to that of the basic alloy. Additions of Cb of more than 1 or 2 percent were somewhat detrimental at 1500° F and had little effect at 1200° F.

It is possible the continued increase in strength from additions of Mo and W in the presence of Cb at 1200° F was associated with the little aging which occurred during tests at that temperature. The tendency for a saturation effect at 1500° F may have been related to the structural instability at the higher temperature.

CONCLUSIONS

The influence of systematic variations of chemical composition on rupture properties at 1500° F was investigated for 62 modifications of a basic alloy containing chromium, nickel, cobalt, iron, molybdenum, tungsten, columbium, carbon, manganese, silicon, and nitrogen. These effects of composition apply only for variation of one element at a time. The effects of simultaneous variations were not studied except for molybdenum, tungsten, and columbium. The actual values for rupture properties apply only to the solution-treated and aged condition studied. It is anticipated that within limits the general effects of composition would apply for other systematically controlled treatments. The following conclusions may be drawn from this investigation:

1. The elements in the basic alloy can be varied individually over quite wide ranges without significantly affecting the rupture properties at 1500° F. Considerable reduction of nickel or cobalt could be tolerated without sacrifice in properties. Substantial reductions of chromium, molybdenum, or tungsten would cause some reduction in strength. Increased carbon up to 0.40 percent would not change rupture strength but would reduce elongation in the rupture test. Increased silicon reduces strength. Increased columbium would be somewhat detrimental to strength. Omission of columbium reduces strength except in the presence of increased amounts of molybdenum and tungsten. All columbium-free alloys have low elongation and large grain size.

2. The 3:2:1 combination of molybdenum, tungsten, and columbium in the basic alloy is the optimum combination of these elements. Lower strengths result from lesser amounts and no improvement from increased amounts. Equal rupture strength to that of the basic alloy can be obtained in columbium-free alloys by increasing the molybdenum plus tungsten to about 8 percent although elongation will be much lower.

3. There is little evidence to indicate that composition differences within the usual commercial limits contribute materially to scatter bands of properties observed in commercial production.

4. Very little can be deduced from the data regarding the fundamental mechanisms by which composition controls properties. Some relationship between carbon, molybdenum, tungsten, and columbium seems to be a

controlling factor in strength. A saturation effect reduces improvement from molybdenum and tungsten in the presence of columbium but not in its absence. The increased elongation in the rupture test resulting from columbium additions appears to be due to its grain refinement, reduction of intergranular precipitation, and influence on general precipitation characteristics.

5. The results are similar to those at 1200° F presented in NACA Report 1058 except that increased chromium was not beneficial at 1500° F and the saturation effect for molybdenum and tungsten did not occur at 1200° F.

University of Michigan
Ann Arbor, Mich., June 1, 1951

REFERENCES

1. Reynolds, E. E., Freeman, J. W., and White, A. E.: Influence of Chemical Composition on Rupture Properties at 1200° F of Forged Chromium-Cobalt-Nickel-Iron Base Alloys in Solution-Treated and Aged Condition. NACA Rep. 1058, 1951. (Supersedes NACA TN 2449.)

2. Frey, D. N., Freeman, J. W., and White, A. E.: Fundamental Effects of Aging on Creep Properties of Solution-Treated Low-Carbon N-155 Alloy. NACA Rep. 1001, 1950. (Supersedes NACA TN 1940.)

TABLE I

CHEMICAL COMPOSITION OF EXPERIMENTAL ALLOYS

Alloy	Aim modification from basic alloy (percent)	Chemical composition (percent) (a)									
		C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	N
-----	Basic analysis	0.15	1.7	0.5	20	20	20	3	2	1	0.12
b_1	Basic	.25	1.54	.40	18.45	19.45	20.35	3.06	2.30	1.05	.11
b_2	Basic	.24	1.62	.53	18.88	17.20	19.74	2.95	2.26	1.11	.12
b_3	Basic	.25	1.68	.60	18.99	18.60	20.64	2.77	2.62	1.12	.15
b_4	Basic	.30	1.87	.74	19.37	19.04	17.95	2.89	2.34	.96	.12
b_5	0.4 C	.55	1.58	.67	17.53	18.24	20.54	2.83	1.75	1.00	.12
b_6	.4 C	.53	1.78	.84	16.27	18.77	18.04	3.32	2.50	.82	.10
c_7	Basic	0.16	1.40	0.66	20.64	17.34	20.52	3.01	2.34	1.11	0.15
c_8	Basic	.14	1.46	.41	21.17	20.16	19.81	3.03	2.08	.96	.14
c_{10}	Basic	.15	1.40	.52	20.89	20.64	19.30	3.05	1.91	1.03	.17
c_{11}	Basic	.16	1.63	.77	21.00	17.13	22.46	3.04	1.97	1.08	.14
c_{12}	Basic	.16	1.61	.67	21.04	18.39	22.00	3.03	2.01	1.15	.14
d_{27}	Basic	.15	----	.58	-----	-----	-----	----	----	----	----
e_{74}	Basic	----	----	----	-----	-----	-----	----	----	----	----
e_{75}	Basic	----	----	----	-----	-----	-----	----	----	----	----
13	0.07 C	0.08	1.80	0.37	20.03	20.84	19.42	3.12	2.10	1.09	----
14	.40 C	.36	----	----	-----	-----	-----	----	----	----	----
15	.40 C	.40	----	----	-----	-----	-----	----	----	----	----
16	.60 C	.57	----	----	-----	-----	-----	----	----	----	----
17	.60 C	.60	1.83	----	18.59	20.25	18.92	----	----	----	----
24	0.03 Mn	0.14	0.03	----	-----	-----	-----	----	----	----	----
19	.25 Mn	.15	.30	0.38	20.07	20.70	19.82	2.99	2.02	1.13	----
20	.50 Mn	.14	.50	----	-----	-----	-----	----	----	----	----
21	1.00 Mn	.12	1.04	----	-----	-----	-----	----	----	----	----
22	2.50 Mn	.14	2.58	----	-----	-----	-----	----	----	----	----
28	1.0 Si	0.13	----	1.19	-----	-----	-----	----	----	----	----
80	1.5 Si	.14	----	1.56	-----	-----	-----	----	----	----	----
51	10 Cr	0.14	----	----	10.18	-----	-----	----	----	----	----
52	30 Cr	.15	----	----	30.51	-----	-----	----	----	----	----
23	0 Ni	0.15	----	----	-----	0.01	-----	----	----	----	----
25	10 Ni	.16	----	----	-----	10.70	-----	----	----	----	----
26	30 Ni	.14	----	----	-----	30.64	-----	----	----	----	----
29	0 Co	0.19	----	----	-----	-----	0.31	----	----	----	----
30	10 Co	.14	----	----	-----	-----	11.09	----	----	----	----
31	30 Co	.14	1.76	0.79	20.50	20.18	32.60	3.41	2.05	1.11	----
32	0 Mo	0.16	----	----	-----	-----	-----	0.20	----	----	----
33	1 Mo	.16	1.76	0.75	20.46	20.52	21.59	1.00	2.02	1.06	0.15
34	2 Mo	.16	----	----	-----	-----	-----	2.34	----	----	----
35	4 Mo	.13	----	----	-----	-----	-----	4.99	----	----	----
36	6 Mo	.12	----	----	-----	-----	-----	6.99	----	----	----
37	0 W	0.14	----	----	-----	-----	-----	----	0.04	----	----
38	1 W	.15	1.74	0.75	20.75	20.50	21.68	3.42	.82	1.02	0.15
39	4 W	.14	----	----	-----	-----	-----	----	4.66	----	----
40	6 W	.16	----	----	-----	-----	-----	----	7.13	----	----
47	0 Cb	0.14	----	----	-----	-----	-----	----	----	0.03	----
48	2 Cb	.14	1.74	0.70	20.65	20.59	20.08	3.20	1.94	1.97	0.14
49	4 Cb	.14	----	----	-----	-----	-----	----	----	4.07	----
50	6 Cb	.16	----	----	-----	-----	-----	----	----	6.09	----
41	Low N	0.14	----	----	-----	-----	-----	----	----	----	0.004
42	0.07 N	.15	----	----	-----	-----	-----	----	----	----	.08
81	.22 N	.15	----	----	-----	-----	-----	----	----	----	.18

[a]Values given only where actual chemical analyses were made; blank spaces indicate same aim value as that of basic alloy.

[b]Alloys 1 to 6 were prepared to develop and standardize melting practice.

[c]Alloys 7, 8, 10, 11, and 12 were prepared to determine control of properties with processing from heat to heat of the basic alloy.

[d]Alloy 27, aimed at low Si, resulted in the basic value.

[e]Deoxidation practice was varied on alloys 74 and 75 of the basic analysis.

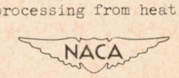


TABLE I.- Concluded

CHEMICAL COMPOSITION OF EXPERIMENTAL ALLOYS

Alloy	Aim modification from basic alloy (percent)			Chemical composition (percent) (a)									
				C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	N
	Mo	W	Cb										
43	0	0	0	0.12	----	----	-----	-----	-----	0.09	0.00	0.00	----
44	0	0	0	.18	1.72	0.74	20.22	20.62	20.09	.00	.00	.00	0.14
45	2	0	0	.14	----	----	-----	-----	-----	2.09	.00	.00	----
46	4	0	0	.13	----	----	-----	-----	-----	4.01	.00	.00	----
53	0	2	0	.13	----	----	-----	-----	-----	.01	2.05	.00	----
54	0	4	0	.14	----	----	-----	-----	-----	.02	3.97	.00	----
55	2	2	0	0.15	1.71	0.72	20.28	20.41	20.20	2.33	2.01	0.00	0.15
56	4	2	0	.13	----	----	-----	-----	-----	4.05	2.04	.00	----
57	2	4	0	.14	----	----	-----	-----	-----	1.98	4.16	.00	----
58	4	4	0	.13	----	----	-----	-----	-----	4.04	4.11	.00	----
59	2	2	2	0.16	----	----	-----	-----	-----	2.19	2.04	1.98	----
60	4	2	2	.13	----	----	-----	-----	-----	4.12	2.07	1.98	----
61	2	4	2	.15	1.69	0.80	19.99	20.50	20.54	2.32	4.12	2.03	0.14
62	2	2	4	.15	----	----	-----	-----	-----	2.02	1.93	4.09	----
63	2	0	2	0.11	----	----	-----	-----	-----	2.10	0.00	1.93	----
64	2	0	4	.15	----	----	-----	-----	-----	2.32	.03	4.23	----
65	4	0	2	.13	----	----	-----	-----	-----	4.02	.00	2.10	----
66	4	0	4	.13	----	----	-----	-----	-----	3.96	.00	4.08	----
67	4	4	4	.13	1.70	0.71	20.18	20.36	20.12	4.12	3.94	3.94	0.17
77	0	0	1	0.15	----	----	-----	-----	-----	----	----	0.96	----
68	0	0	2	.18	----	----	-----	-----	-----	0.00	0.00	1.81	----
69	0	0	4	.14	----	----	-----	-----	-----	.00	.00	3.76	----
70	0	2	2	.16	----	----	-----	-----	-----	.01	2.00	1.92	----
71	0	4	2	.17	----	----	-----	-----	-----	.02	3.93	1.91	----
72	0	2	4	.14	1.80	0.66	20.21	20.39	20.51	.01	2.01	4.18	0.12
73	0	4	4	.15	----	----	-----	-----	-----	.02	3.92	4.05	----
82	2	4	4	0.16	----	----	-----	-----	-----	----	----	----	----
83	4	2	4	.13	----	----	-----	-----	-----	3.85	1.85	4.38	----
84	4	4	2	.14	1.64	0.74	20.18	20.51	19.95	----	----	2.25	0.17

[a]Values given only where actual chemical analyses were made; blank spaces indicate same aim value as that of basic alloy.

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TABLE II

RUPTURE TEST CHARACTERISTICS AT 1500° F FOR MODIFIED ALLOYS

[Treatment: 2200° F, 1 hr, water-quenched; 1400°F, 24 hr, air-cooled]

Alloy	Alloy modification (percent)	Stress (psi)	Rupture time (hr)	Elongation in 1 in. (percent)	Reduction of area (percent)	Minimum creep rate (percent/hr)	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent in 1 in.)	Stress (psi) to cause creep rates of -		Minimum creep rate at 13,000 psi (percent/hr)
							100 hr	1000 hr		0.01 percent/hr	0.1 percent/hr	
8	Basic	19,000	47	18	20.3	------	16,500	11,800	16	13,900	18,000	0.0056
		17,000	105	14	17.6	0.056						
		15,000	125	15	13.8	.043						
		12,000	[a]446	--	4.0	.0038						
		11,000	[b]1649	--	----	.0010						
10	Basic	20,000	56	37	47.0	------	18,500	12,900	30	15,000	18,100	.0028
		19,000	98	34	37.8	------						
		18,000	98	27	32.7	.090						
		16,500	275	19	20.3	.015						
		15,000	446	17	23.2	.010						
		13,000	684	6	4.8	.0029						
		12,000	[c]1537	--	----	------						
12	Basic	19,000	53	41	34	------	17,100	11,900	30	13,500	17,500	.0076
		17,000	106	26	30.0	.096						
		14,000	365	16	16.0	.0112						
		12,000	904	15	16.8	.0044						
27	Basic	18,000	58	35	46.0	------	16,500	12,000	32	12,800	16,400	.0120
		15,000	160	29	34.6	.050						
		13,500	424	36	41.0	.0165						
		12,000	1086	34	37.2	.0068						
74	Basic, no deoxidation	20,000	53	41	32.7	------	18,400	12,500	43	13,400	18,500	.0078
		18,000	130	46	48.9	.082						
		15,000	317	28	42.1	.023						
		14,000	515	15	14.4	.013						
75	Basic, Zr-Si-Fe deoxidant	20,000	55	45	48.1	-----	18,400	12,500	28	14,700	18,500	.0028
		18,000	106	28	42.3	.083						
		15,000	416	29	34.2	.012						
		13,000	837	20	25.4	-----						
13	0.08 C	18,000	65	34	39.0	-----	16,400	[d]10,700	38	11,400	15,500	0.0275
		15,000	190	47	56.7	0.075						
		13,000	328	43	52	.0275						
15	.40 C	18,000	[a]89	--	11.0	.05	17,900	[d]13,700	4	16,000	19,000	.00080
		16,500	182	4	5.6	.0099						
		15,000	407	4	4	.005						
16	.60 C	18,000	48	13	14.7	.192	15,500	10,300	8	12,500	16,500	.0130
		15,000	113	6	13.8	.049						
		12,000	494	10	10.9	.0064						
		10,000	1234	6	12.5	------						
24	0 Mn	20,000	33	6	10.2	------	17,000	13,000	5	15,000	[d]20,000	0.0041
		18,000	82	2	5.6	0.0315						
		15,000	223	8	6.4	.012						
		14,000	291	5	5.6	.0046						
		13,000	1008	21	25.2	.0041						
19	.30 Mn	19,000	41	16	9.4	------	16,000	10,300	12	14,500	[d]19,500	.0046
		15,000	161	8	5.5	.013						
		13,000	356	5	4.0	.0046						
		11,000	662	6	4	.0018						
		10,000	1324	5	2	------						
20	.50 Mn	19,000	50	16	13.8	.133	17,000	11,500	10	13,800	18,500	.007
		17,000	108	9	8.6	.049						
		15,000	227	11	10.9	.0185						
		13,000	460	10	8.7	.007						
		11,500	1186	7	8	------						
21	1.0 Mn	19,000	33	28	26.6	------	16,400	11,200	20	13,200	17,200	.0095
		17,000	84	24	21.8	.111						
		15,000	200	12	16.0	.020						
		13,000	342	12	11.7	.014						
		11,500	827	19	10.3	------						
22	2.5 Mn	19,000	57	29	36.5	------	17,100	12,400	26	14,000	18,500	.005
		17,000	105	26	25.9	.083						
		15,000	208	14	18.3	.018						
		13,000	726	14	17.6	.005						
28	1.2 Si	17,000	58	23	25.2	------	15,300	10,500	20	13,300	15,200	0.0070
		15,000	122	17	16.0	0.07						
		12,500	303	7	6.4	.0031						
		10,000	1394	10	5.6	------						
80	1.6 Si	15,000	74	34	30.7	.126	14,000	11,000	33	11,800	13,900	.044
		13,000	187	28	25.2	.044						
		11,500	719	10	17.6	.0053						

[a]Piece missing.
[b]Overheated.
[c]Adapter broke.
[d]Estimated.



TABLE II. - Continued

RUPTURE TEST CHARACTERISTICS AT 1500° F FOR MODIFIED ALLOYS

Alloy	Alloy modification (percent)	Stress (psi)	Rupture time (hr)	Elongation in 1 in. (percent)	Reduction of area (percent)	Minimum creep rate (percent/hr)	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent in 1 in.)	Stress (psi) to cause creep rates of -		Minimum creep rate at 13,000 psi (percent/hr)
							100 hr	1000 hr		0.01 percent/hr	0.1 percent/hr	
51	10 Cr	17,000	36	26	28	------	15,000	[d]10,300	30	11,200	15,300	0.030
		16,000	77	20	19.0	0.109						
		15,000	86	30	29.5	.109						
		13,500	229	22	24.6	.0305						
		12,000	318	27	27.5	.021						
52	30 Cr	20,000	18	23	61.6	------	17,500	[d]12,000	30	14,500	20,000	.0045
		18,000	79	42	55.8	.047						
		16,000	175	18	50.5	.021						
		14,000	363	13	32.7	.0083						
23	0 Ni	17,000	25	41	39.0	------	13,200	9,400	21	10,300	13,300	0.039
		16,000	23	32	39.4	------						
		15,000	81	25	33.5	0.11						
		13,500	81	23	32.3	.11						
		12,000	272	17	21.8	.0245						
		10,500	286	25	24.6	.019						
		9,000	1007	15	13.8	------						
25	10 Ni	20,000	51	36	43.8	------	18,500	13,100	33	15,100	18,300	.0015
		18,000	128	32	44.5	.078						
		15,500	516	30	42.6	.007						
		15,000	493	32	37.2	.011						
		13,500	791	32	41.8	------						
26	30 Ni	18,000	38	19	18.9	------	15,800	11,000	13	13,000	16,100	.010
		16,000	90	13	13.8	.084						
		15,000	102	13	13.0	.052						
		12,500	594	10	8.6	.0063						
		11,500	428	7	7.1	------						
29	0 Co	18,000	60	14	27.5	------	15,800	9,200	19	13,500	17,000	0.0067
		15,000	110	19	25.9	0.0249						
		14,000	181	13	12.4	.015						
		12,500	330	8	9.3	.0039						
		10,000	618	4	3	------						
30	10 Co	19,500	49	44	47.9	------	17,500	[d]12,000	28	14,700	18,700	.0029
		18,000	78	29	43.3	.107						
		15,000	262	22	34.2	.012						
		13,500	446	16	28.0	.0041						
31	32 Co	18,000	67	14	18.3	.13	16,500	10,200	15	11,800	17,100	.0175
		15,000	153	15	14.4	.040						
		13,000	409	13	13.0	.015						
		11,000	562	12	11	.0070						
		10,000	1160	10	8	------						
32	0 Mo	17,000	67	50	42.1	------	15,600	11,000	45	[d]11,800	14,400	0.030
		15,000	122	43	49.1	0.096						
		12,500	539	27	30.0	.035						
		11,000	944	32	30.1	------						
33	1 Mo	17,000	50	23	23.2	------	15,500	11,000	20	12,300	15,000	.018
		15,000	127	19	21.1	.081						
		13,000	272	27	30.0	.024						
		11,500	730	10	14.4	.0023						
34	2 Mo	17,000	74	28	33.4	.161	16,400	[d]12,500	25	13,100	16,000	.0090
		16,000	105	25	24.8	.135						
		14,100	364	27	26.2	.023						
35	5 Mo	22,000	10	26	30.1	------	17,300	12,900	20	15,200	18,700	.0015
		18,000	92	20	24.8	.045						
		16,000	156	17	25.9	.021						
		13,000	992	10	12.4	.0015						
36	7 Mo	18,000	76	54	50.4	.085	16,700	11,000	45	13,000	18,500	.0094
		15,000	182	34	44.5	.026						
		13,000	397	33	37.2	.0094						
		10,500	1279	11	28.8	------						
37	0 W	17,000	54	50	47.3	------	15,400	10,200	40	10,500	15,000	0.0360
		15,000	134	32	41.8	0.099						
		13,500	292	36	40.3	.035						
		11,500	524	33	36.5	.0160						
38	1 W	18,000	40	30	29.5	------	15,100	[d]9,800	25	10,400	16,000	.034
		16,500	62	22	27.3	.186						
		15,000	132	27	30.0	.077						
		13,000	235	25	25.2	.034						
		11,500	434	20	21.1	.0170						

[d]Estimated.

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TABLE II. - Continued

RUPTURE TEST CHARACTERISTICS AT 1500° F FOR MODIFIED ALLOYS

Alloy	Alloy modification (percent)			Stress (psi)	Rupture time (hr)	Elongation in 1 in. (percent)	Reduction of area (percent)	Minimum creep rate (percent/hr)	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent in 1 in.)	Stress (psi) to cause creep rates of -		Minimum creep rate at 13,000 psi (percent/hr)
									100 hr	1000 hr		0.01 percent/hr	0.1 percent/hr	
39	5 W			21,000	44	44	52.6	------	18,800	13,000	36	15,100	20,000	0.0028
				18,000	166	29	50.4	0.045						
				16,000	[b]279	--	----	.015						
				15,500	397	33	39.0	.010						
				13,500	727	28	38	------						
40	7 W			21,000	26	50	56.5	------	17,100	[d]12,100	44	14,600	18,000	.0027
				20,000	39	43	47.9	------						
				18,000	144	45	54.3	.096						
				16,500	168	51	61.6	.040						
				16,500	209	41	58.5	.031						
				13,500	(b)	--	----	.003						
47	0 Cb			15,000	40	7	6.4	------	12,700	[d]8,400	6	11,200	14,000	0.046
				13,500	70	7	4.8	0.066						
				12,000	148	5	4.7	.019						
				10,000	389	2	2.4	.0036						
48	2 Cb			18,000	37	47	53.8	------	16,000	11,000	45	13,100	16,000	.0092
				15,500	125	43	61.4	.079						
				14,000	312	48	58.9	.021						
				12,000	634	29	48	------						
49	4 Cb			17,000	42	74	76.4	------	15,100	11,300	50	12,000	14,500	.021
				15,000	110	51	73.2	.149						
				13,000	526	53	71.0	.021						
				12,000	557	48	72.0	.018						
50	6 Cb			20,000	26	48	54.9	------	16,000	10,800	45	12,000	14,500	.026
				17,000	91	39	66.2	.165						
				15,000	156	56	61.4	.155						
				12,000	592	38	49.7	.0075						
41	0.004 N			18,000	59	45	61.0	------	16,700	11,800	43	13,000	15,600	0.010
				17,000	91	43	59.8	0.186						
				15,000	206	37	52.6	.070						
				13,000	545	36	53.6	.01						
42	.08 N			19,000	65	28	27.3	-----	17,600	12,400	27	13,900	17,300	.0050
				17,000	124	27	34.0	.068						
				15,000	310	29	33.5	.0194						
				13,500	533	16	16.0	.0076						
81	.18 N			20,000	46	37	37.8	------	17,600	[d]12,400	38	[d]13,300	17,000	.0080
				18,000	100	38	34.6	.162						
				16,000	170	35	45.1	.057						
				15,000	276	31	37.8	.034						
	Mo	W	Cb											
43	0	0	0	10,000	64	7	5.5	0.065	9,100	4,500	8	7,700	10,000	[d]0.80
				8,500	135	10	5.6	.042						
				8,000	166	6	3.2	.0115						
				6,000	390	4	.7	.00025						
				5,000	711	6	1.6	.0002						
45	2	0	0	13,000	49	9	6.4	.083	11,300	[d]6,800	7	9,800	13,100	.083
				12,000	95	6	4.8	.038						
				11,500	99	7	5.6	.0315						
				10,000	144	3	4.0	.0125						
				8,500	385	4	2.4	.0028						
46	4	0	0	15,000	39	7	5.6	------	12,500	8,400	5	11,300	[d]14,800	.033
				12,000	128	3	3.2	.016						
				10,000	334	7	1.6	.0035						
				8,000	1941	4	7.1	------						
53	0	2	0	12,000	42	6	6.4	------	9,500	5,100	6	8,000	11,600	.20
				10,000	83	7	4.0	.037						
				8,000	152	2	.8	.0085						
				6,000	547	4	.8	.002						
54	0	4	0	12,000	66	4	4.8	.036	11,000	6,700	4	9,800	13,100	.092
				10,000	178	5	4.8	.012						
				8,500	328	8	1.6	.0038						
				8,000	506	4	1.6	.0016						
				6,500	1179	6	1.6	-----						

[b]Overheated.
[d]Estimated.

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TABLE II. - Continued

RUPTURE TEST CHARACTERISTICS AT 1500° F FOR MODIFIED ALLOYS

Alloy	Alloy modification (percent)			Stress (psi)	Rupture time (hr)	Elongation in 1 in. (percent)	Reduction of area (percent)	Minimum creep rate (percent/hr)	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent in 1 in.)	Stress (psi) to cause creep rates of -		Minimum creep rate at 13,000 psi (percent/hr)
									100 hr	1000 hr		0.01 percent/hr	0.1 percent/hr	
	Mo	W	Cb											
55	2	2	0	14,000	54	6	3.2	0.05	12,700	6,800	5	12,000	15,500	0.021
				12,000	152	4	2.4	.0075						
				10,500	276	4	2.4	.0035						
				10,000	347	3	1.6	.0018						
				8,000	624	4	1	------						
56	4	2	0	18,000	54	12	11.0	.161	16,000	9,500	8	13,800	17,500	.0055
				15,000	131	6	5.6	.022						
				12,000	422	4	3.2	.0025						
				10,000	673	2	1.6	------						
57	2	4	0	15,000	69	5	4.8	.045	14,100	8,700	7	12,800	16,600	.011
				13,500	117	8	2.4	.018						
				12,000	304	5	1.6	.005						
				11,000	422	4	1.6	.0025						
				9,000	880	10	.8	------						
58	4	4	0	20,000	30	10	7.9	------	17,400	11,800	9	15,000	18,400	.0019
				18,000	89	9	7.9	.063						
				15,000	240	7	4.0	.0114						
				13,000	551	4	2.4	.0019						
59	2	2	2	18,000	54	61	55.3	------	16,100	10,800	50	[d]11,200	15,200	.032
				15,000	148	37	46.8	.091						
				13,000	369	40	44.5	.032						
				10,500	1162	27	31.5	------						
60	4	2	2	18,000	51	48	66.7	------	17,000	12,500	58	14,900	17,000	.0010
				17,000	105	58	65.8	.092						
				15,000	280	39	55.3	.02						
				14,000	540	38	48.8	.0033						
				13,000	558	30	48	------						
61	2	4	2	19,000	55	33	57.3	------	17,300	11,500	45	13,400	16,500	.0062
				18,000	75	42	57.0	------						
				16,000	223	59	55.8	.060						
				14,000	364	39	50.0	.0139						
				12,000	531	21	34	------						
62	2	2	4	18,000	40	76	64.6	------	15,500	[d]11,000	60	[d]10,700	14,600	.042
				15,000	130	51	66.2	.113						
				13,000	334	62	65.9	.042						
63	2	0	2	18,000	46	81	69.5	------	15,500	[d]10,900	60	-------	14,400	[d]0.044
				16,500	70	64	68.9	0.40						
				15,000	247	56	53.0	.066						
				13,500	300	48	68.8	.055						
				12,000	400	54	64.2	.040						
64	2	0	4	16,500	45	70	70.4	------	14,400	10,900	70	10,200	13,800	.061
				15,000	85	73	70.3	.194						
				13,000	237	58	63.9	.061						
				11,500	724	49	61.9	.0235						
65	4	0	2	19,000	43	59	58	------	16,900	10,200	27	13,000	17,400	.010
				17,000	102	27	46.2	.087						
				15,000	267	42	56.8	.028						
				13,000	541	35	41.1	.011						
				11,500	746	34	37.9	------						
66	4	0	4	20,000	23	84	68.5	------	15,200	[d]10,600	50	11,300	13,800	.054
				18,000	42	65	65.5	------						
				14,000	135	43	65.0	.122						
				12,000	432	32	63.3	.018						
67	4	4	4	18,000	52	52	62.5	------	16,000	10,800	46	11,800	15,500	.024
				16,000	105	46	65.8	.110						
				13,000	245	42	61.4	.030						
				12,000	545	45	59.4	.011						
77	0	0	1	15,000	77	30	38.4	0.209	14,300	9,100	29	10,400	14,000	0.057
				12,500	226	20	28.2	.037						
				10,000	758	19	18.3	.0078						
68	0	0	2	14,500	42	46	45.8	------	12,500	8,200	31	[d]8,600	12,000	.173
				13,000	82	31	33.5	.173						
				11,500	170	32	36.5	.08						
				10,000	365	21	27.3	.027						
				8,500	910	8	13.1	------						
69	0	0	4	15,000	47	75	63.5	------	13,500	[d]9,100	60	[d]9,100	12,500	.125
				13,500	82	65	63.5	------						
				12,000	233	42	48.6	.070						
				10,500	470	35	41.8	.0300						

[d]Estimated.



TABLE II. - Concluded

RUPTURE TEST CHARACTERISTICS AT 1500° F FOR MODIFIED ALLOYS

Alloy	Alloy modification (percent)			Stress (psi)	Rupture time (hr)	Elongation in 1 in. (percent)	Reduction of area (percent)	Minimum creep rate (percent/hr)	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent in 1 in.)	Stress (psi) to cause creep rates of -		Minimum creep rate at 13,000 psi (percent/hr)
									100 hr	1000 hr		0.01 percent/hr	0.1 percent/hr	
	Mo	W	Cb											
70	0	2	2	15,000	69	25	34.8	-----	14,800	[d]12,100	26	12,400	15,000	0.017
				14,000	244	28	31.3	0.05						
				13,000	440	15	24.6	.017						
71	0	4	2	18,000	44	47	48.9	-----	16,200	[d]12,200	25	12,500	15,100	.014
				17,000	72	34	47.3	-----						
				16,000	95	25	37.9	.20						
				15,000	252	42	42.9	.057						
				13,500	400	32	40.3	.025						
72	0	2	4	17,000	41	76	70.6	-----	14,500	d9,800	65	[d]11,300	13,200	.079
				17,000	32	46	61.9	-----						
				15,000	89	68	55.8	.183						
				13,000	200	46	57.0	.079						
73	0	4	4	18,000	34	55	65.4	-----	14,700	11,900	56	11,900	14,000	.035
				16,500	22	69	61.0	-----						
				15,000	97	56	65.0	.197						
				13,500	268	51	59.4	.063						
				12,000	945	42	41.0	.012						
82	2	4	4	18,000	42	69	71.4	-----	16,000	11,500	67	11,500	15,300	0.024
				16,000	99	67	73.0	0.153						
				14,000	225	53	73.1	.050						
				12,500	590	71	73.1	.021						
				11,500	455	63	56.8	-----						
83	4	2	4	18,000	37	66	70.6	-----	15,300	10,700	44	11,500	15,300	.024
				16,500	66	42	72.1	.174						
				15,000	161	46	68.9	.095						
				13,000	219	51	70.0	.020						
				9,500	[e]2156	--	----	.0018						
84	4	4	2	18,000	64	43	58.3	------	17,300	12,600	47	14,400	19,000	.0043
				17,000	112	50	57.6	.087						
				15,000	358	52	63.8	.014						
				13,500	579	40	57.3	.0056						

[d]Estimated.

[e]Turned off.

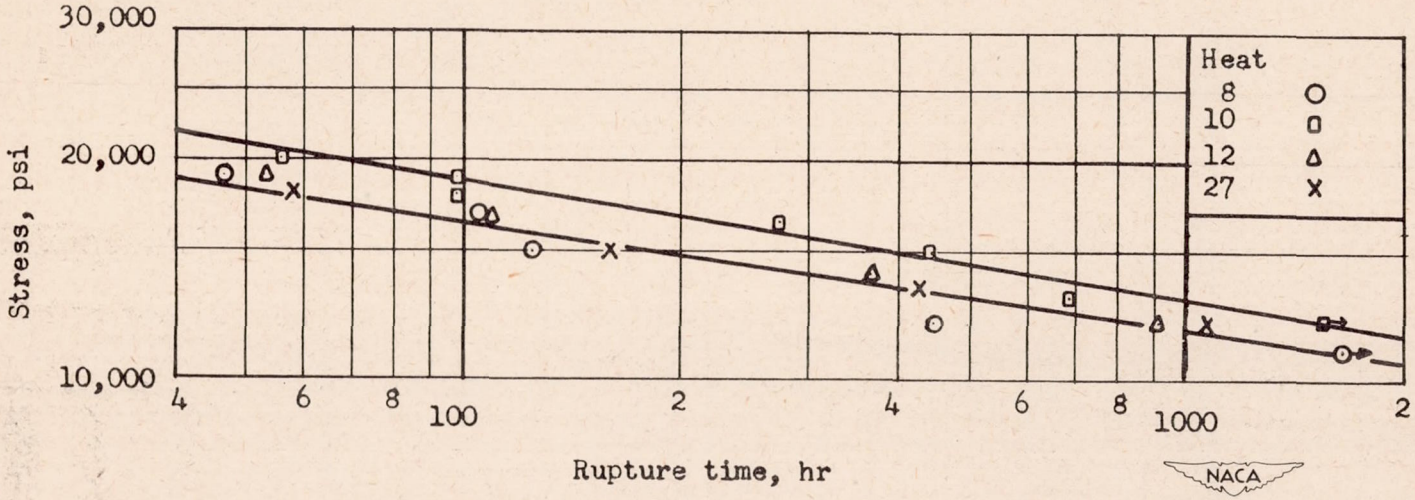


Figure 1.- Curves of stress against time for rupture at 1500° F for four heats of basic alloy. Treatment: 2200° F for 1 hour, water-quenched; 1400° F for 24 hours, air-cooled.

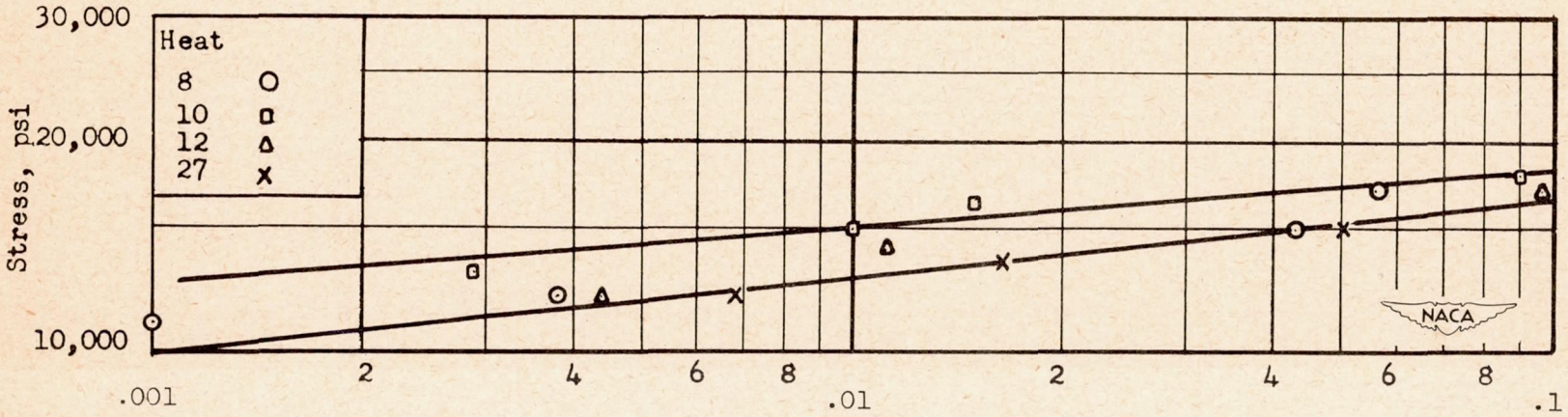


Figure 2.- Curves of stress against minimum creep rate at 1500° F for four heats of basic alloy. Treatment: 2200° F for 1 hour, water-quenched; 1400° F for 24 hours, air-cooled.

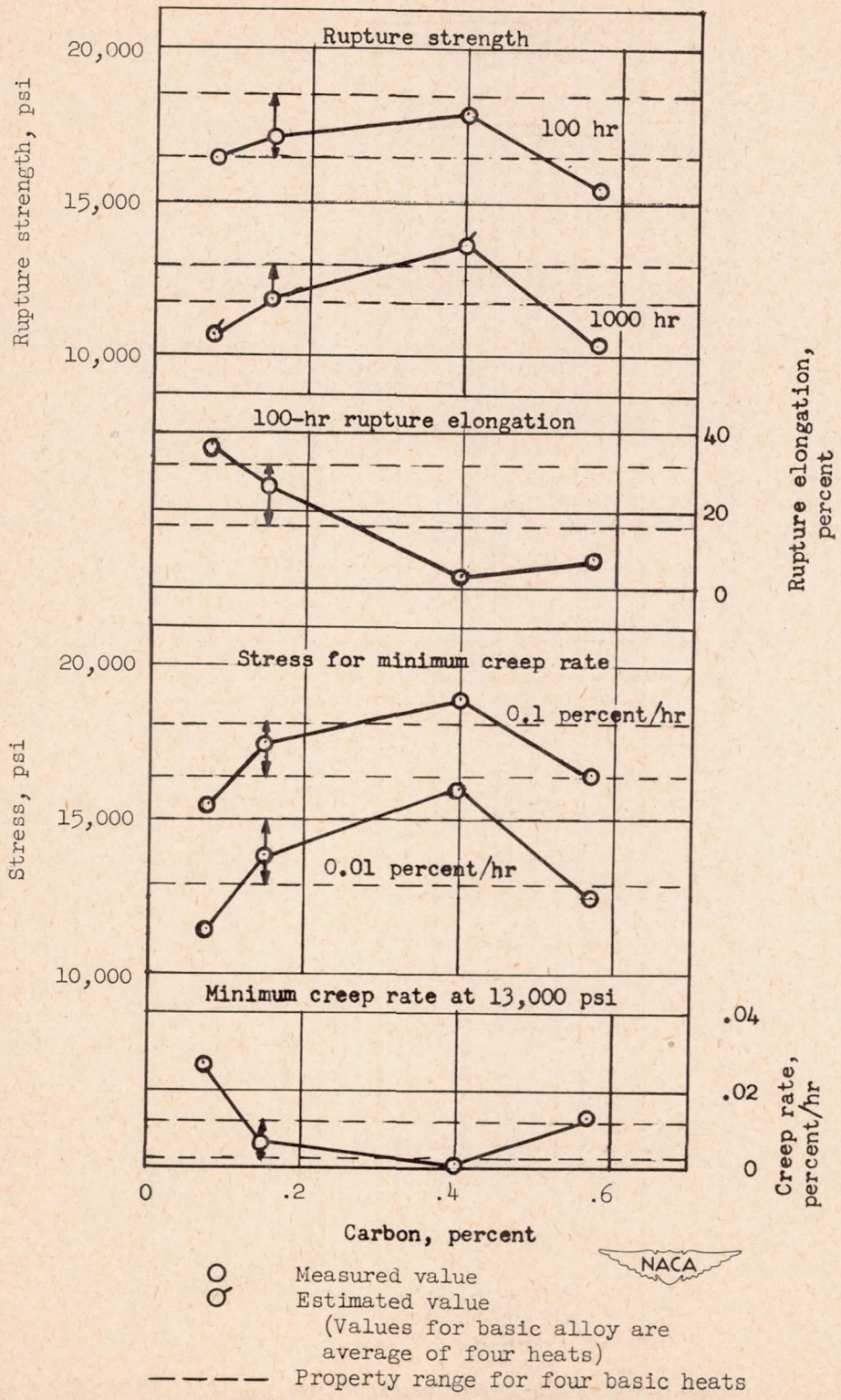


Figure 3.- Curves of rupture test data at 1500° F against carbon content of basic alloy.

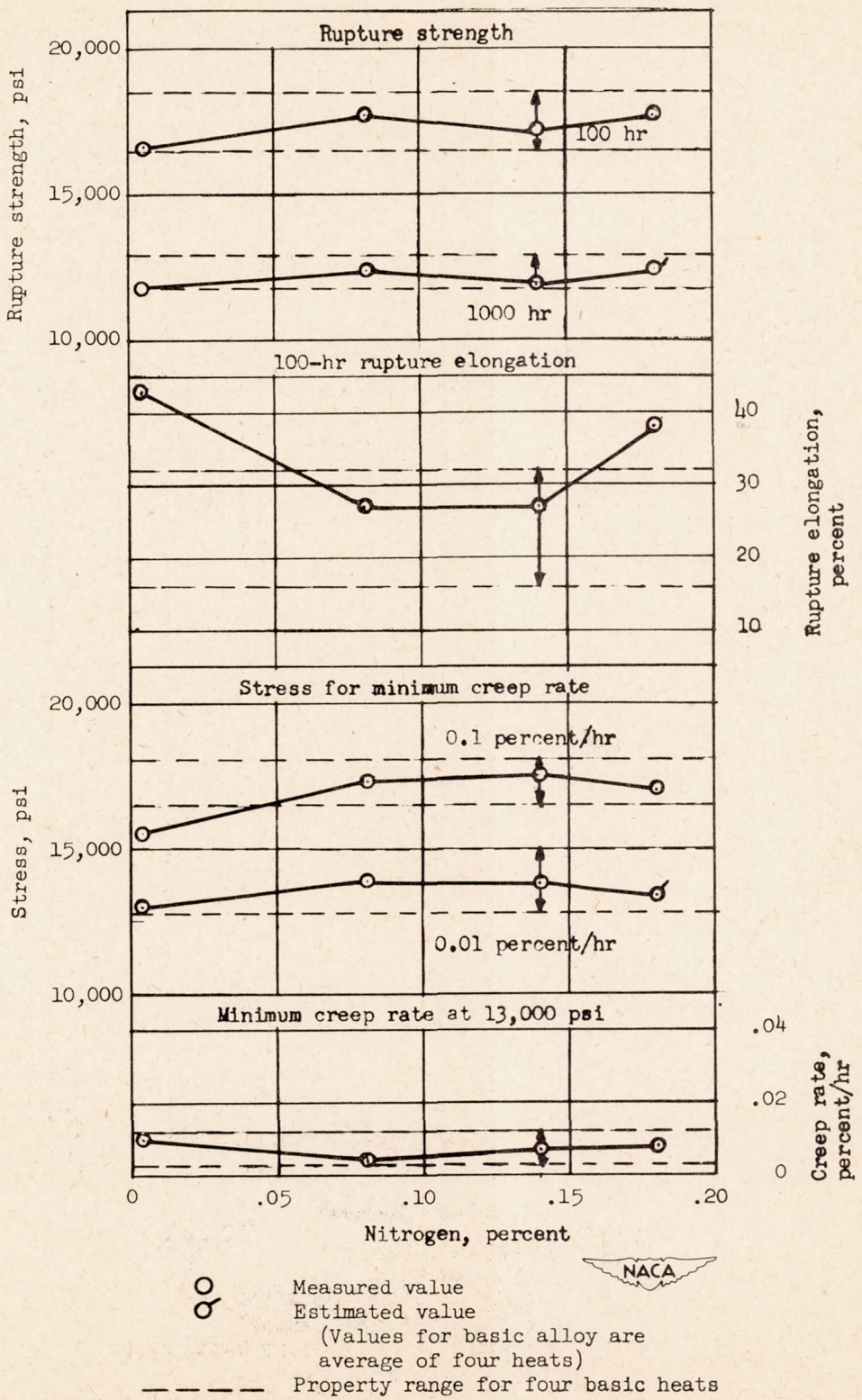


Figure 4.- Curves of rupture test data at 1500° F against nitrogen content of basic alloy.

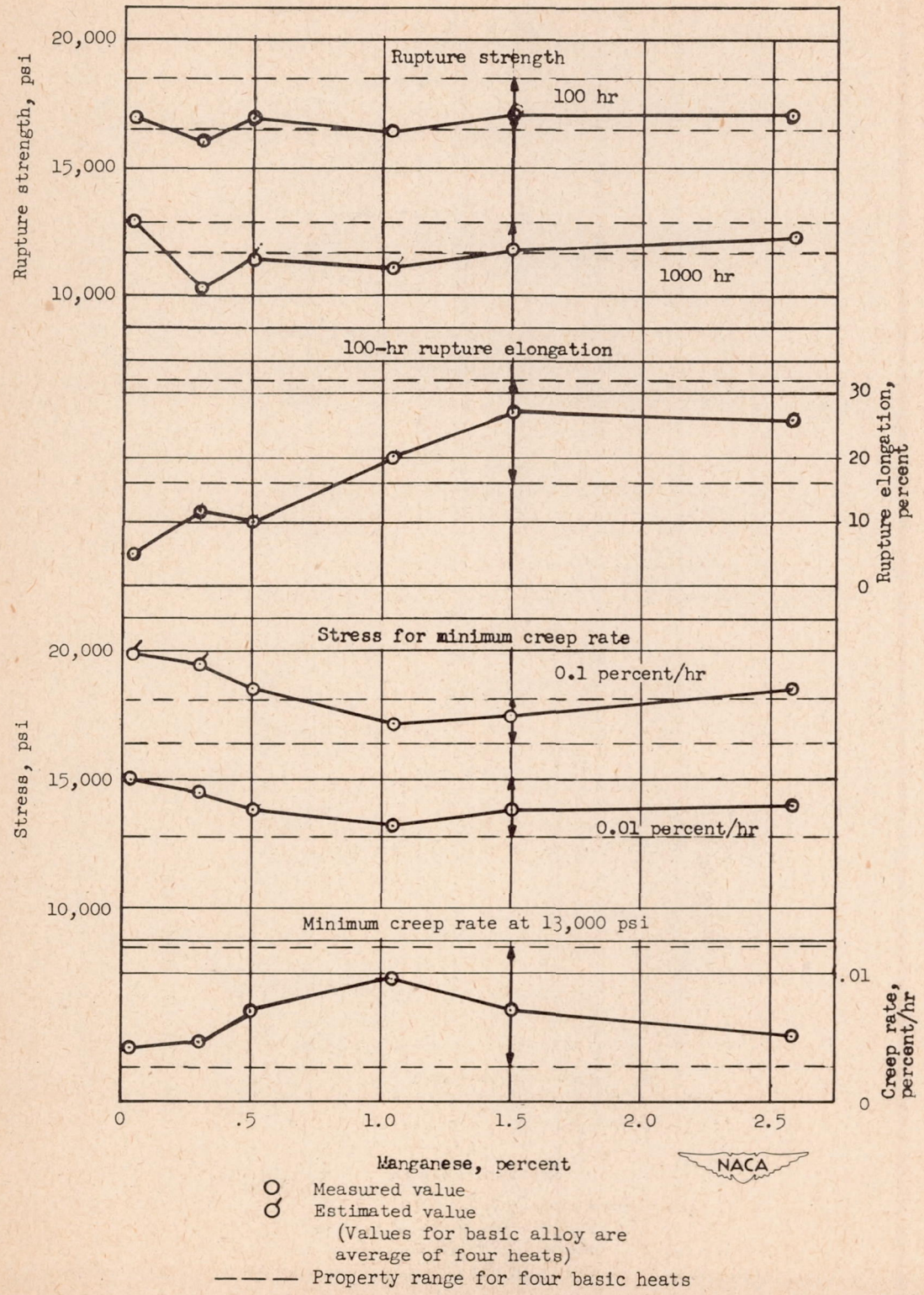


Figure 5.- Curves of rupture test data at 1500° F against manganese content of basic alloy.

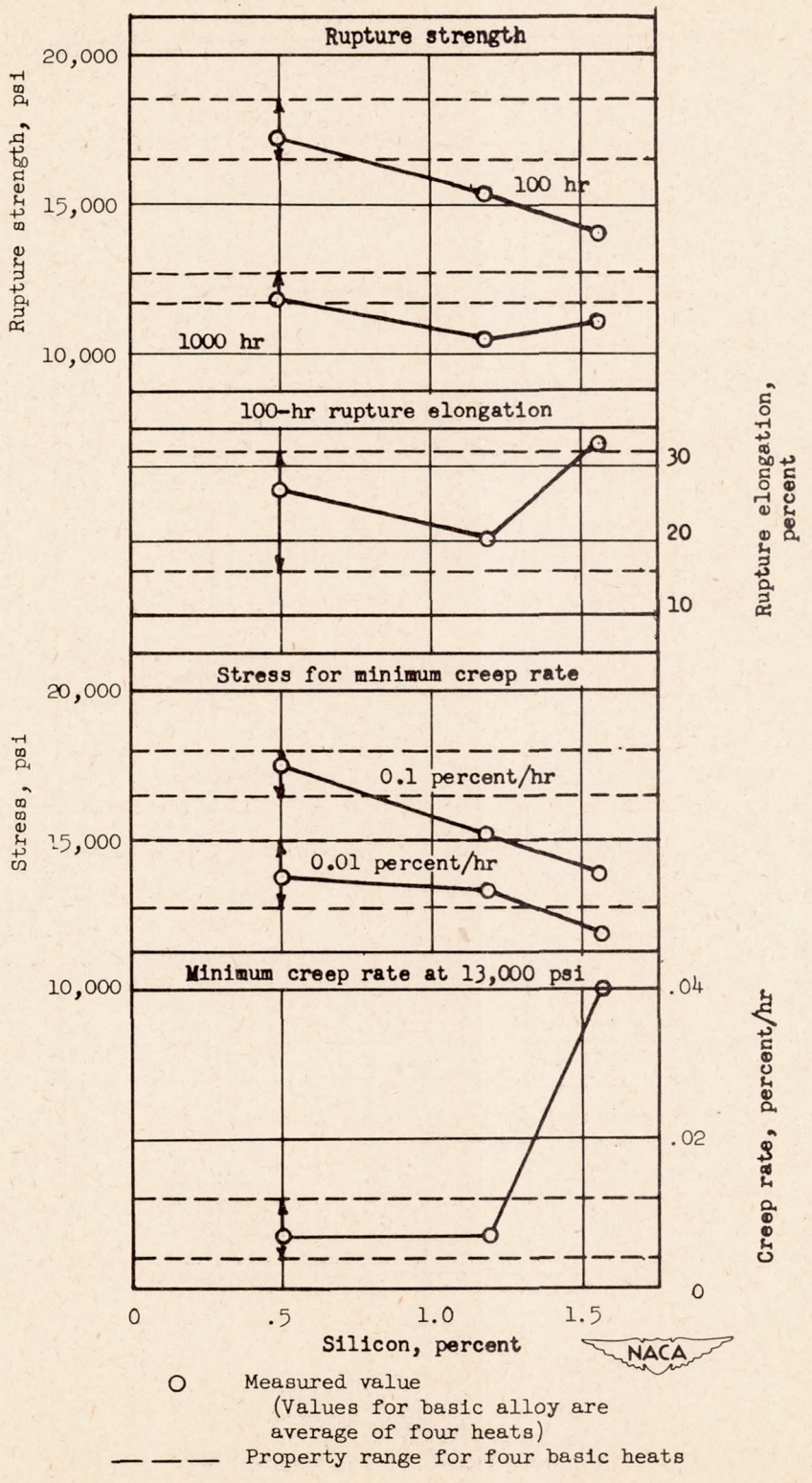


Figure 6.- Curves of rupture test data at 1500° F against silicon content of basic alloy.

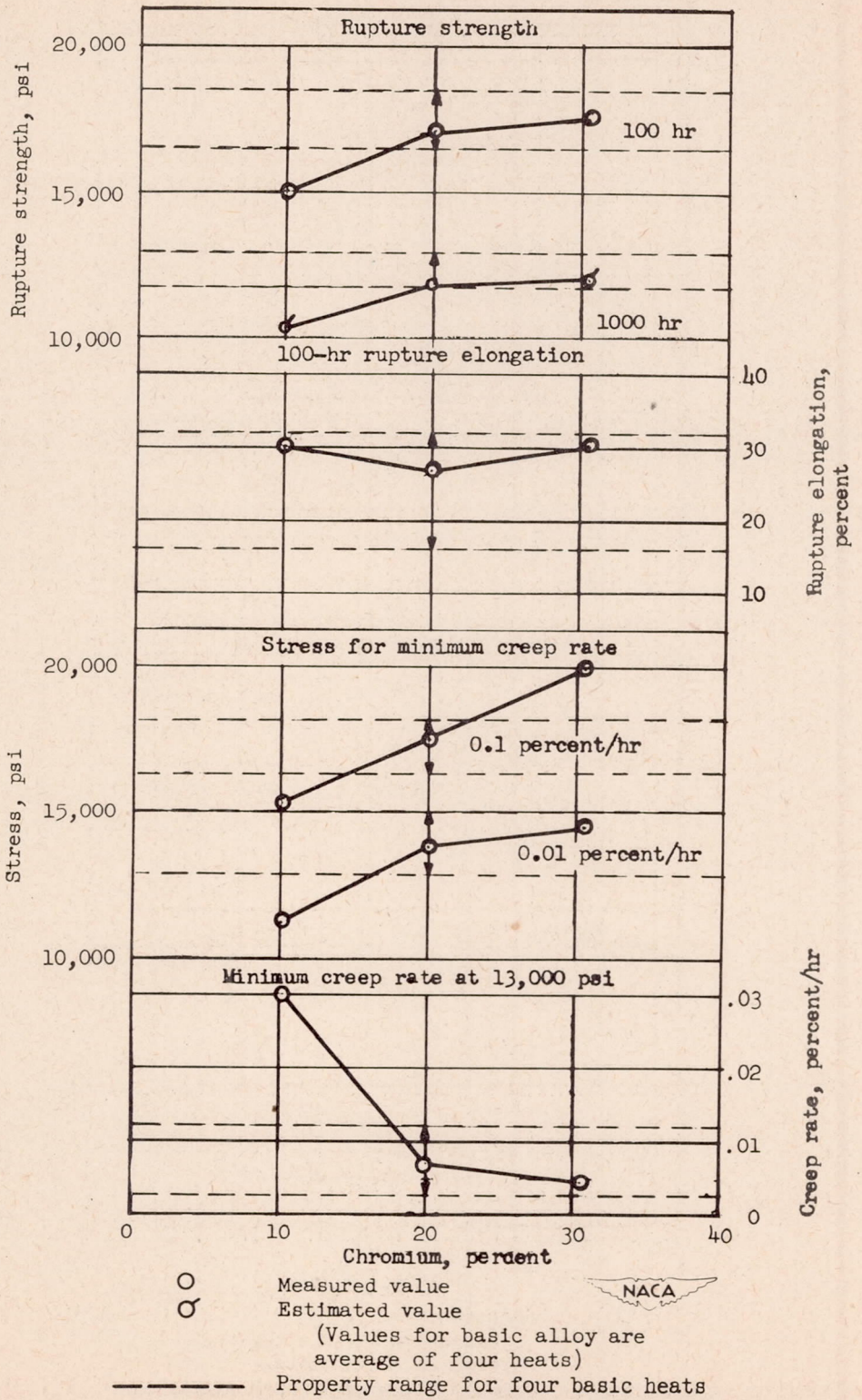


Figure 7.- Curves of rupture test data at 1500° F against chromium content of basic alloy.

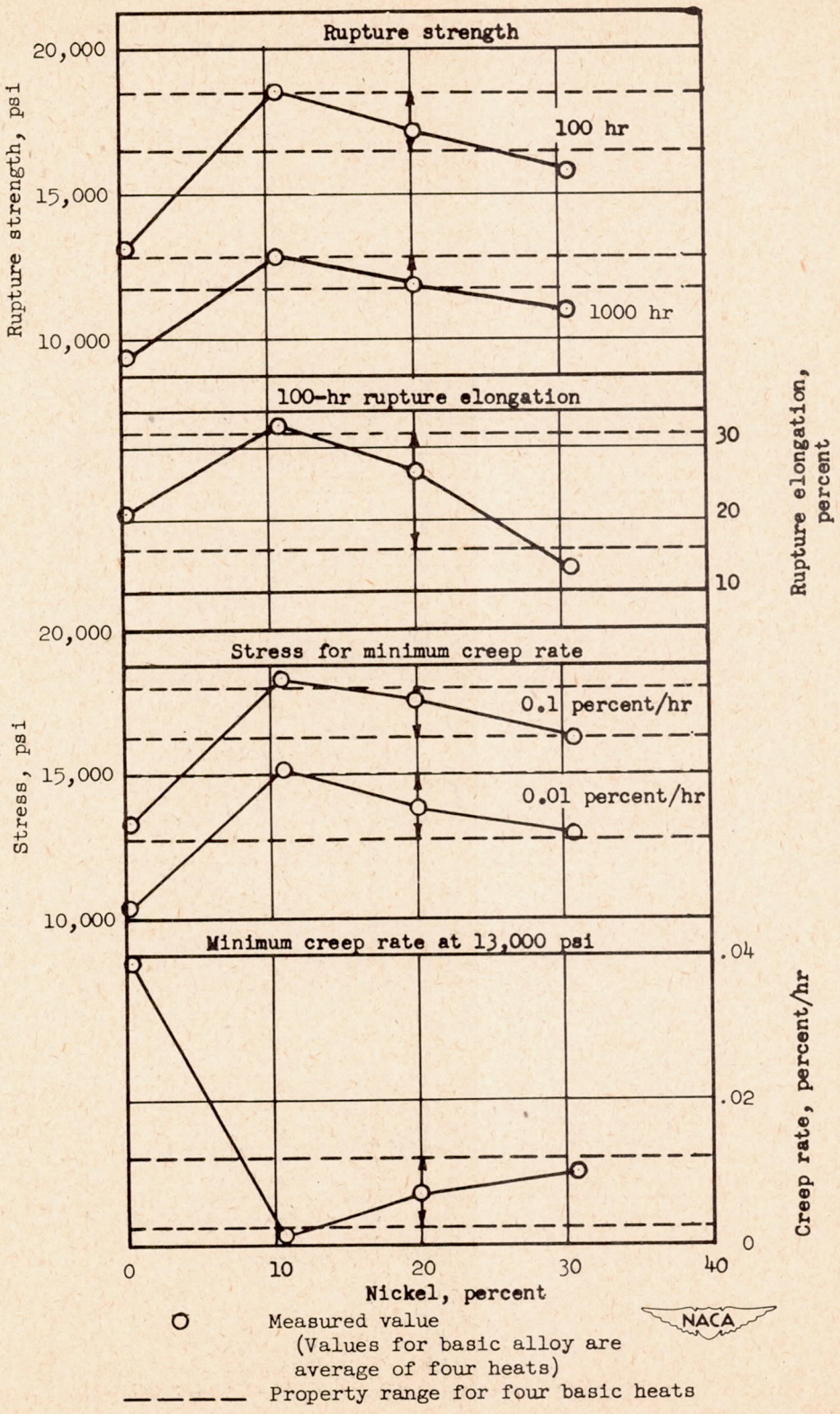


Figure 8.- Curves of rupture test data at 1500° F against nickel content of basic alloy.

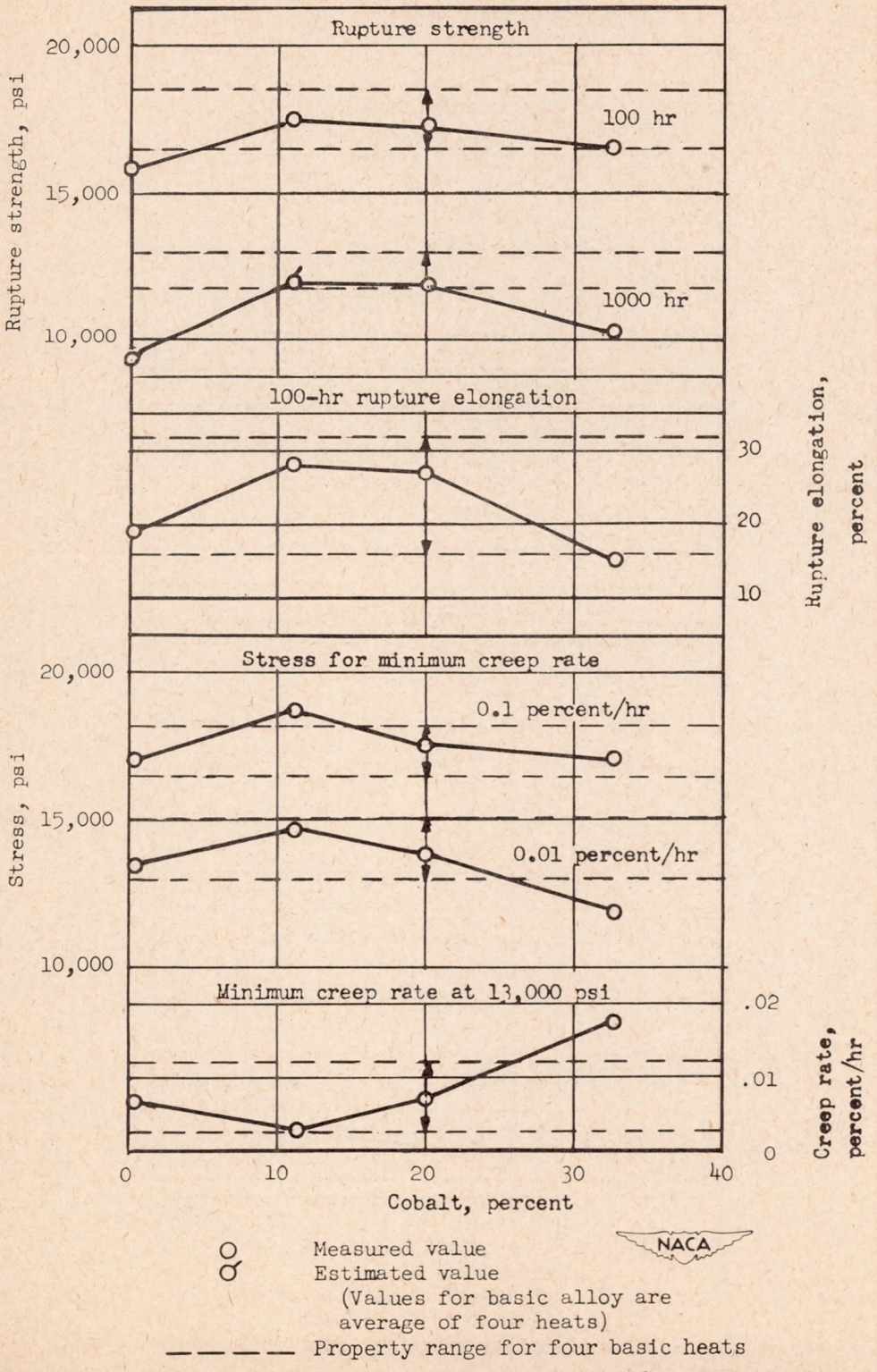


Figure 9.- Curves of rupture test data at 1500° F against cobalt content of basic alloy.

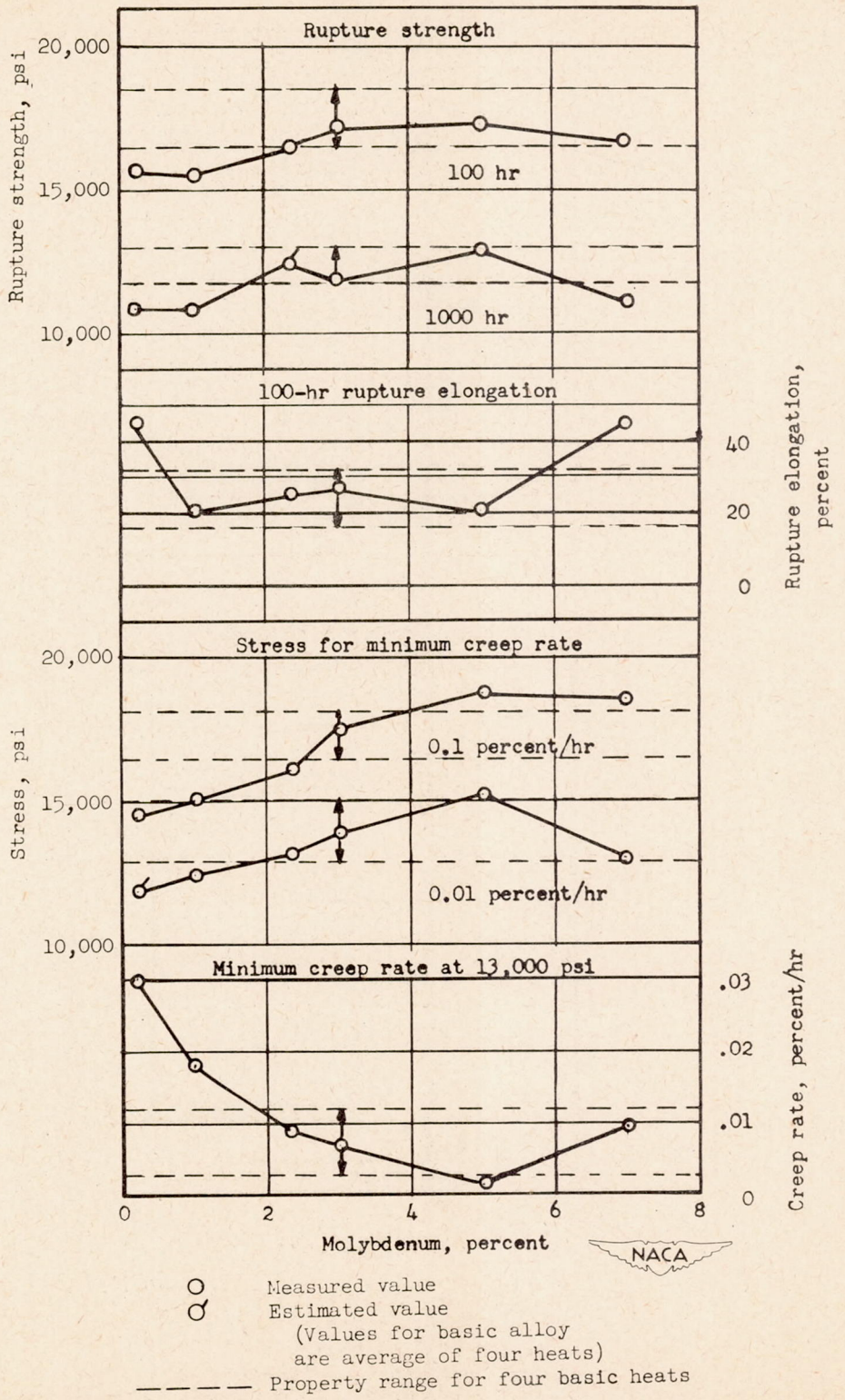


Figure 10.- Curves of rupture test data at 1500° F against molybdenum content of basic alloy.

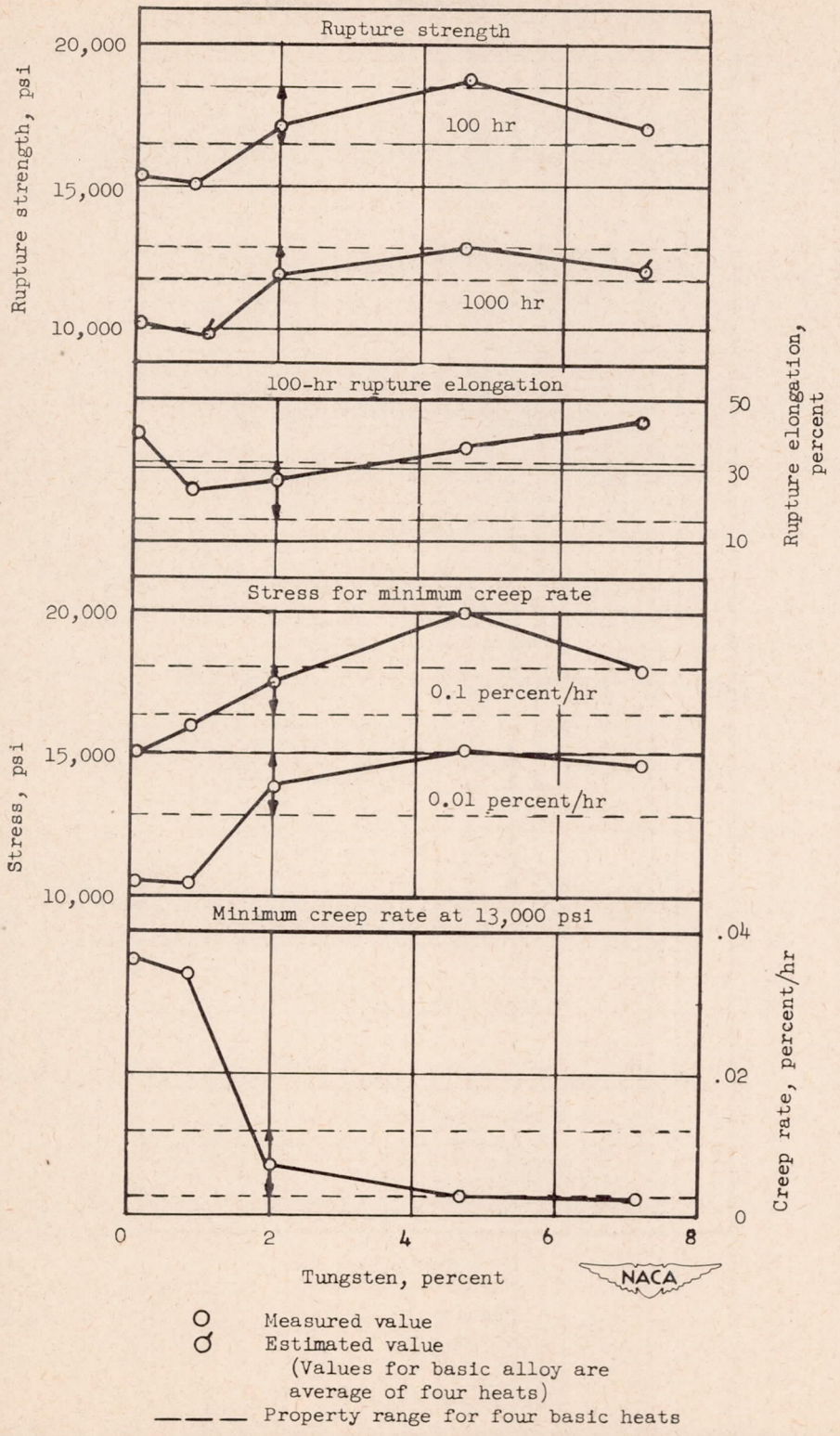


Figure 11.- Curves of rupture test data at 1500° F against tungsten content of basic alloy.

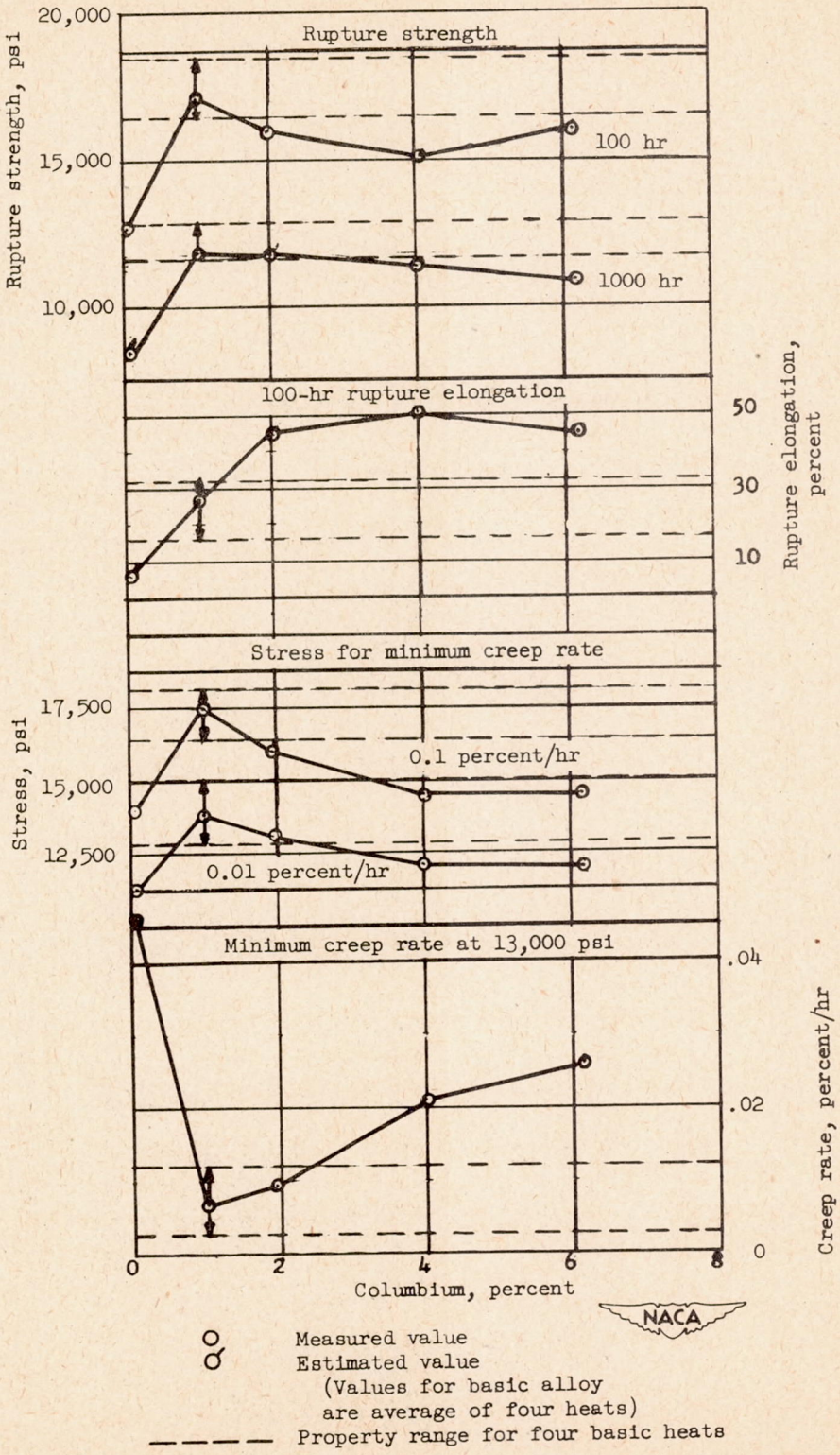


Figure 12.- Curves of rupture test data at 1500° F against columbium content of basic alloy.

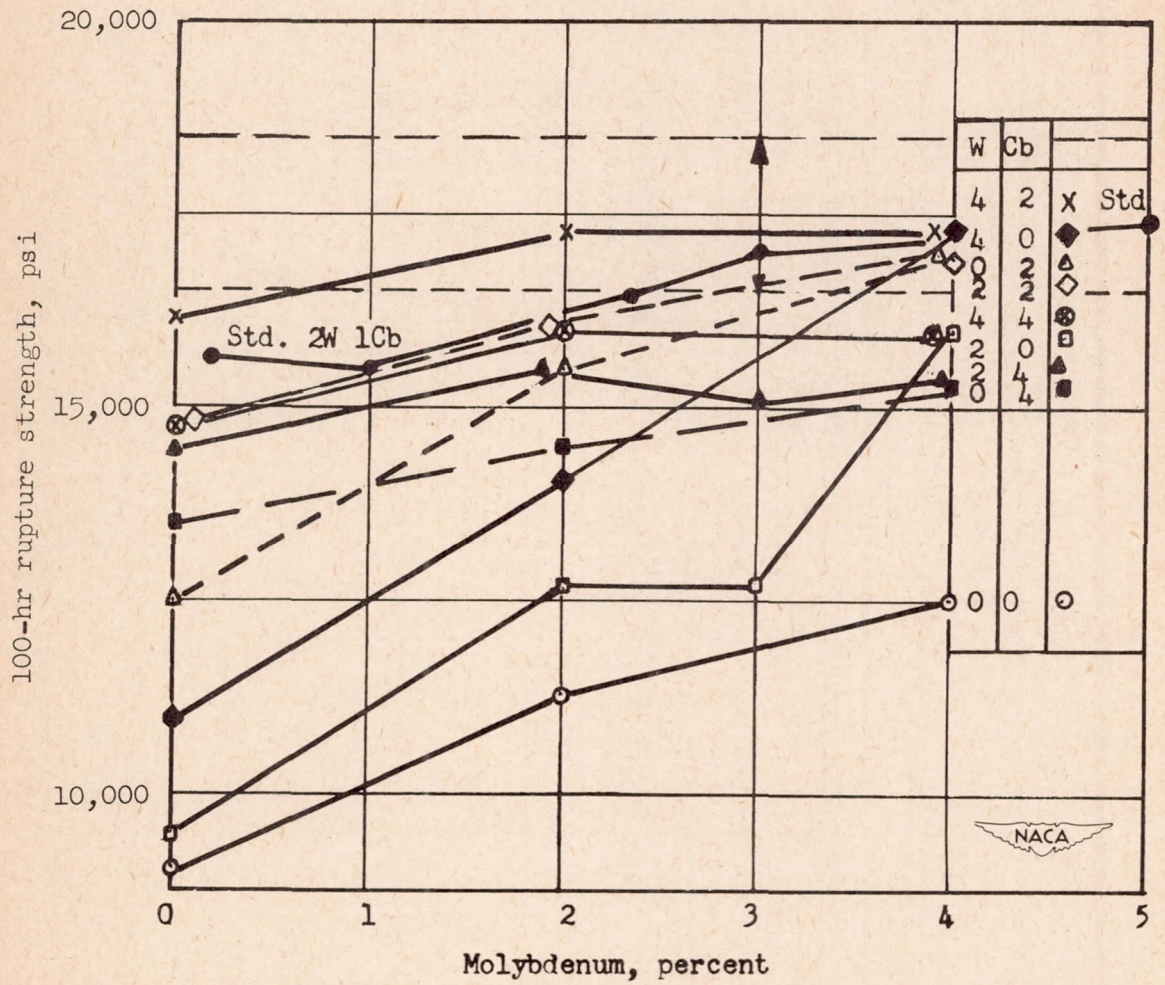


Figure 13.- Summary of influence of molybdenum on 100-hour rupture strength at 1500° F of 10 tungsten-columbium modifications. Horizontal dashed lines indicate property range for four basic heats.

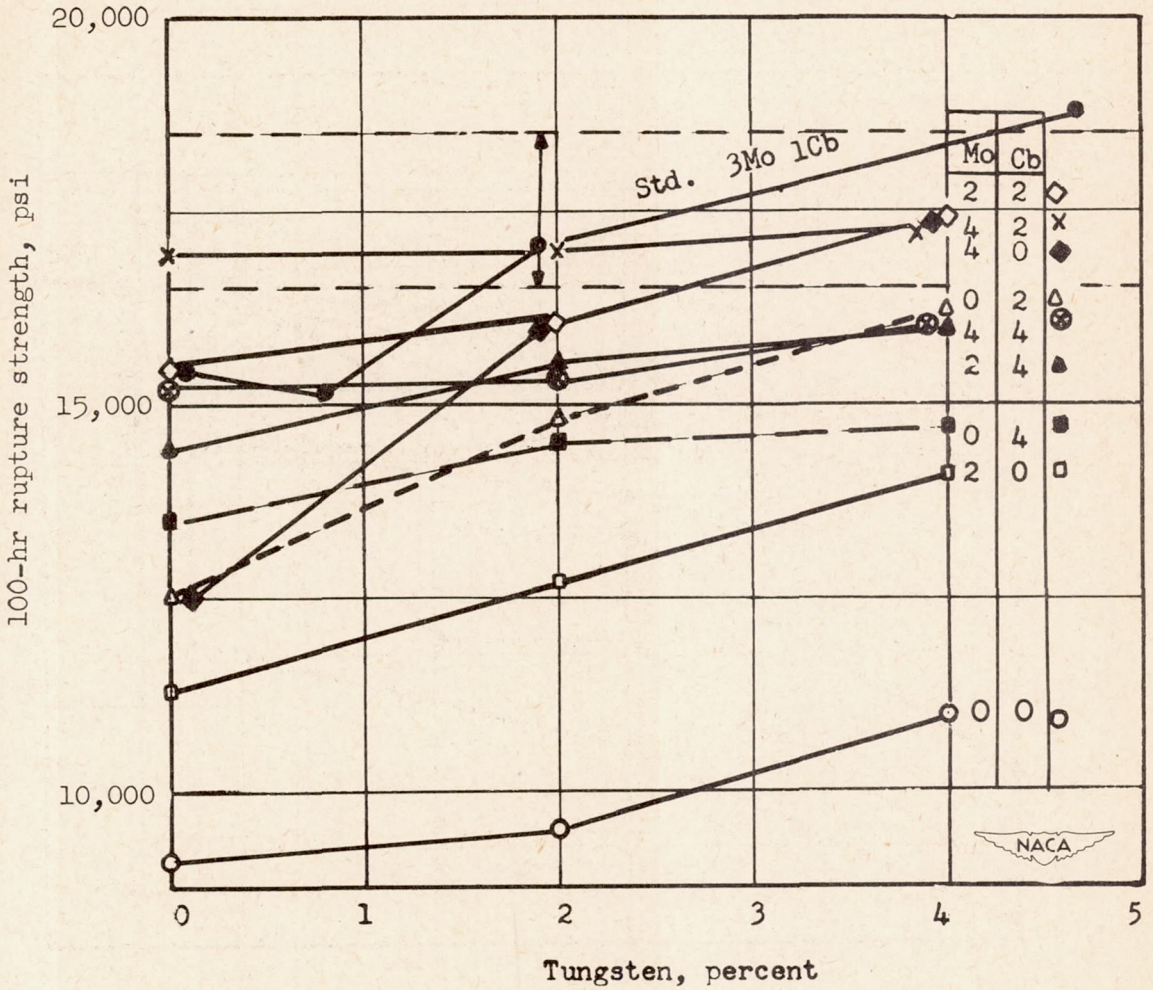


Figure 14.- Summary of influence of tungsten on 100-hour rupture strength at 1500° F of 10 molybdenum-columbium modifications. Horizontal dashed lines indicate property range for four basic heats.

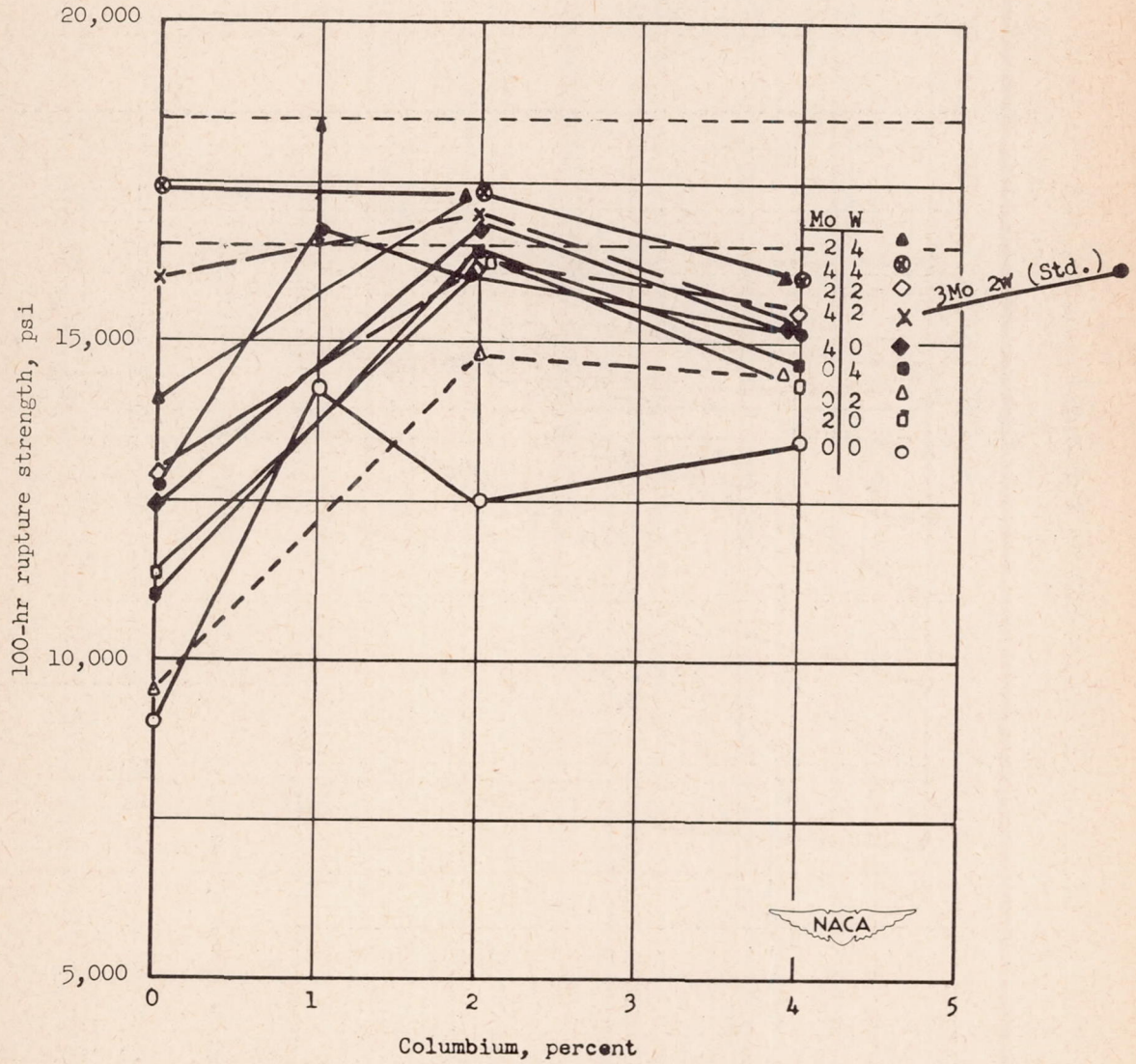


Figure 15.- Summary of influence of columbium on 100-hour rupture strength at 1500° F of 10 molybdenum-tungsten modifications. Horizontal dashed lines indicate property range for four basic heats.

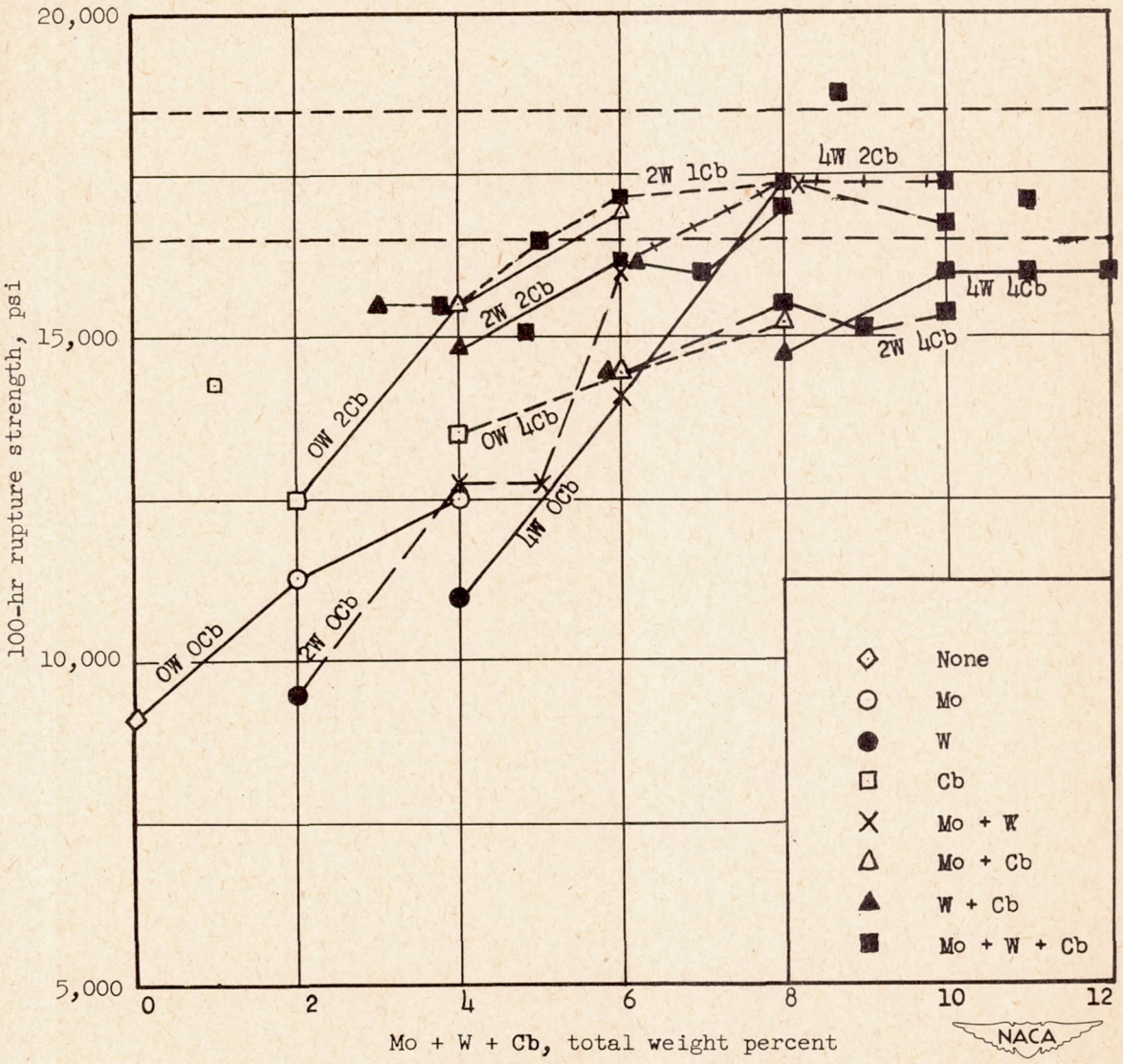


Figure 16.- Influence of molybdenum additions (0, 2, and 4 percent) on a total weight percent basis of molybdenum, tungsten, and columbium on 100-hour rupture strength at 1500° F of modified alloys. Horizontal dashed lines indicate property range for four basic heats.

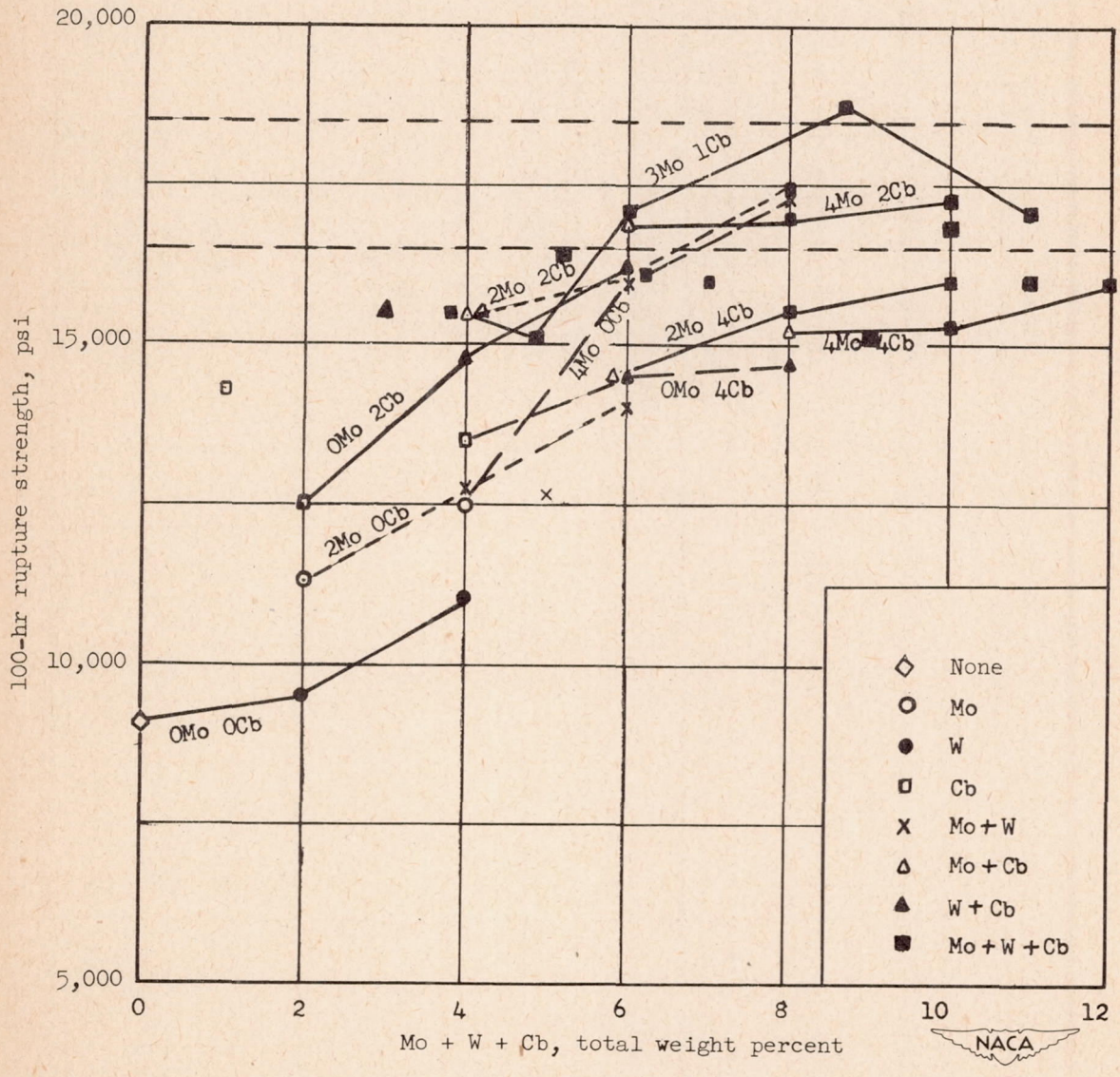


Figure 17.- Influence of tungsten additions (0, 2, and 4 percent) on a total weight percent basis of molybdenum, tungsten, and columbium on 100-hour rupture strength at 1500° F of modified alloys. Horizontal dashed lines indicate property range for four basic heats.

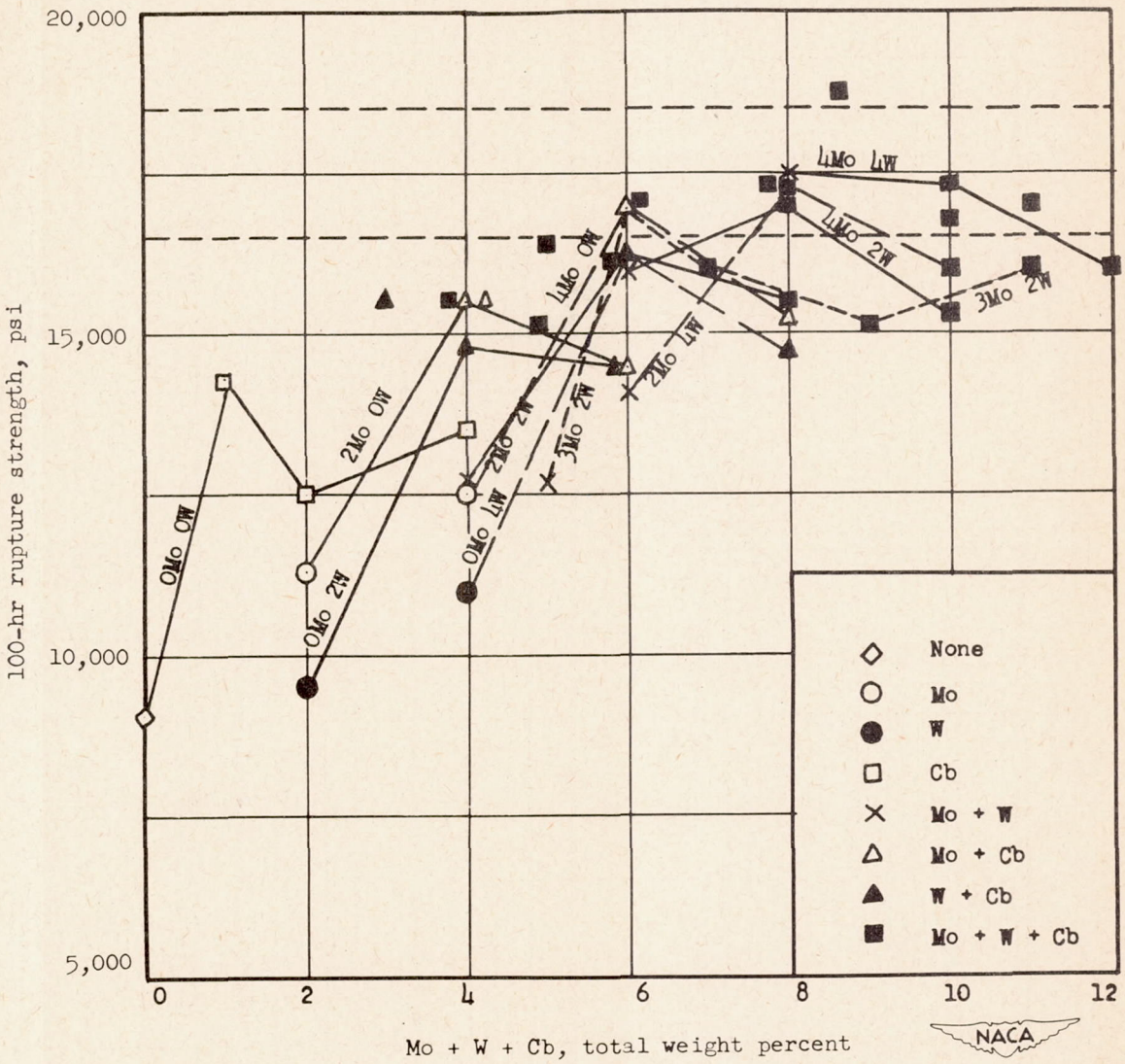


Figure 18.- Influence of columbium additions (0, 2, and 4 percent) on a total weight percent basis of molybdenum, tungsten, and columbium on 100-hour rupture strength at 1500° F of modified alloys. Horizontal dashed lines indicate property range for four basic heats.

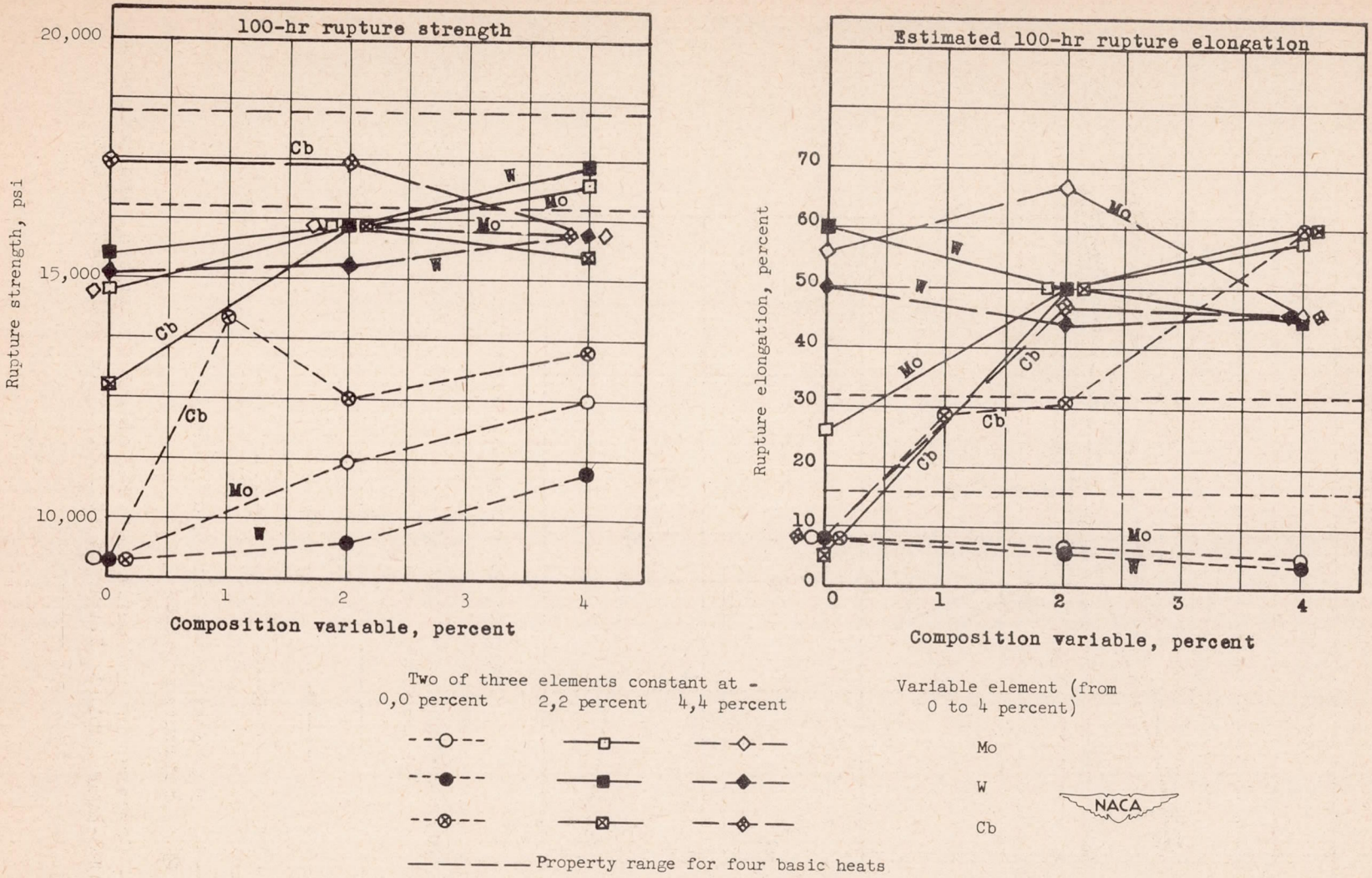
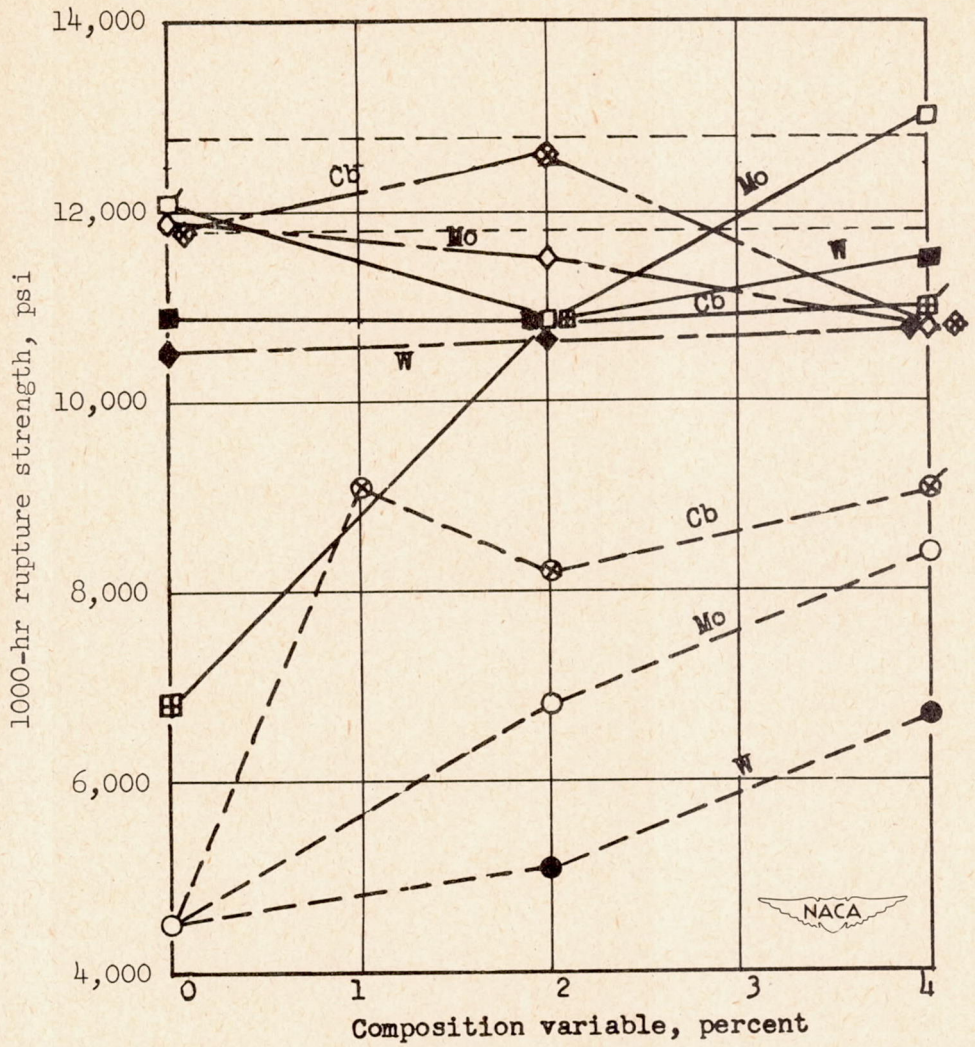


Figure 19.- Comparative influences of molybdenum, tungsten, and columbium on 100-hour rupture properties at 1500° F of modified alloys.



Two of three elements constant at -

0,0 percent	2,2 percent	4,4 percent	Variable element (from 0 to 4 percent)
--○--	—□—	—◇—	Mo
--●--	—■—	—◆—	W
--⊗--	—⊠—	—⊗—	Cb

— — — Property range for four basic heats

□ ⊠ ⊗ Estimated values

Figure 20.- Comparative influences of molybdenum, tungsten, and columbium on 1000-hour rupture strength at 1500° F of modified alloys.

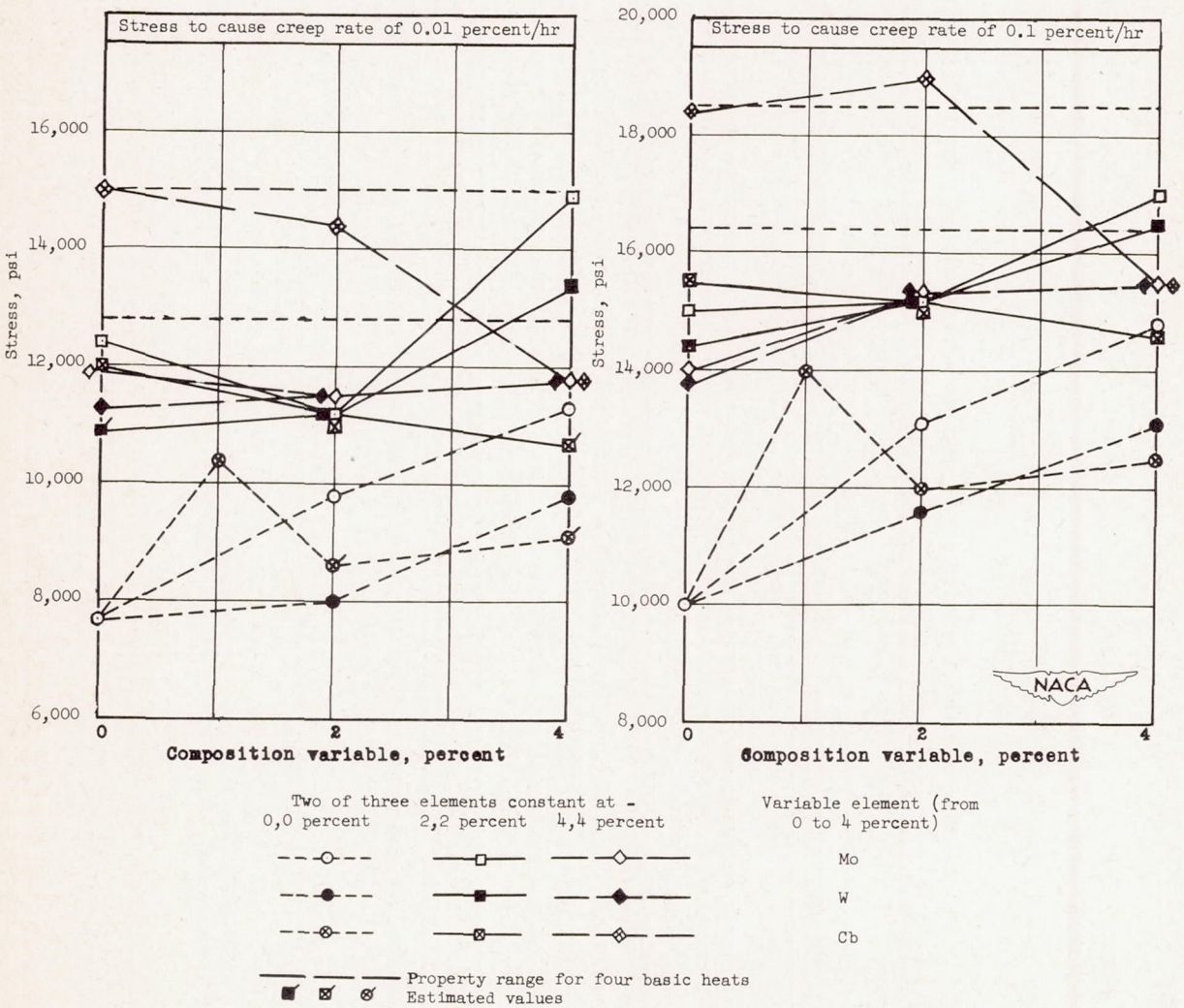


Figure 21.- Comparative influences of molybdenum, tungsten, and columbium on 0.01- and 0.1-percent-per-hour creep strengths at 1500° F of modified alloys.

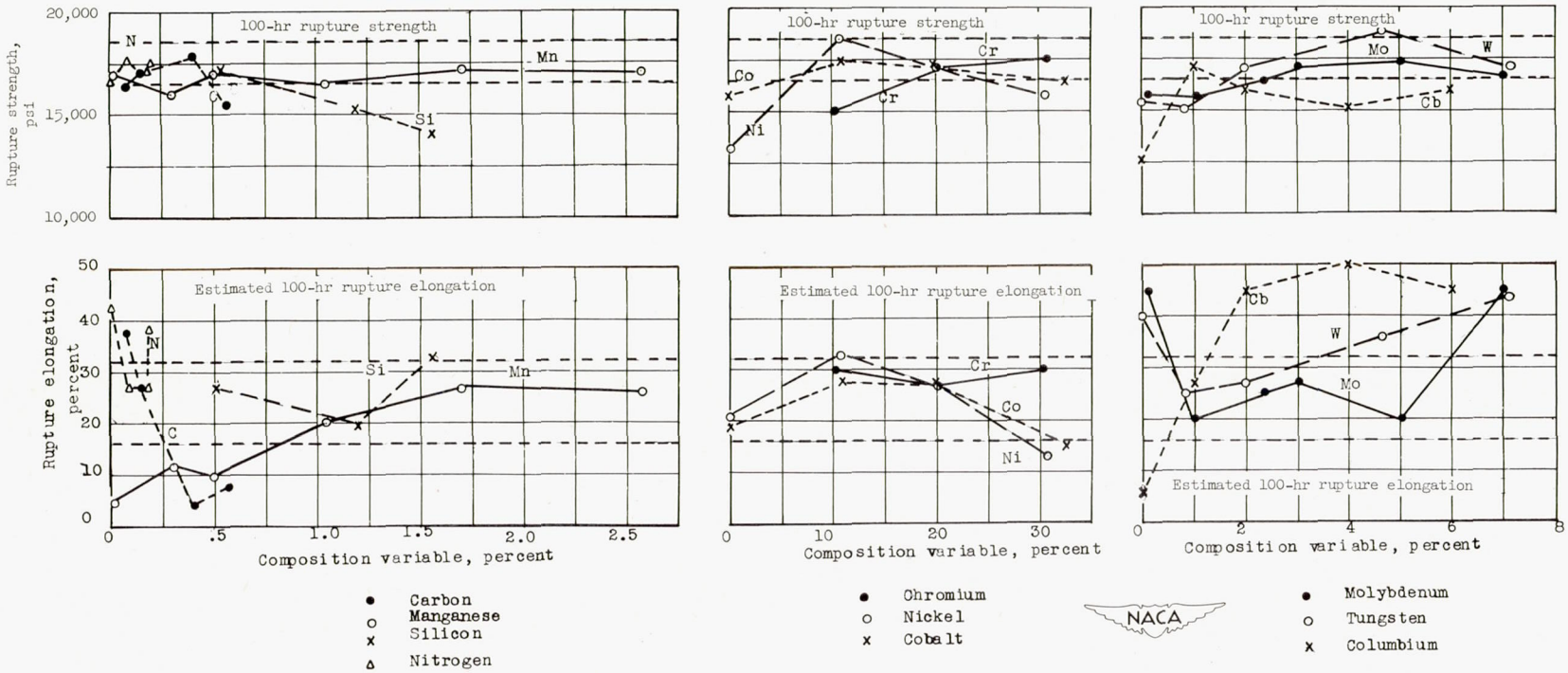


Figure 22.- Summary of influence of variation of individual alloying elements in basic alloy on 100-hour rupture strengths and elongations at 1500° F. Horizontal dashed lines indicate property ranges for four basic heats.

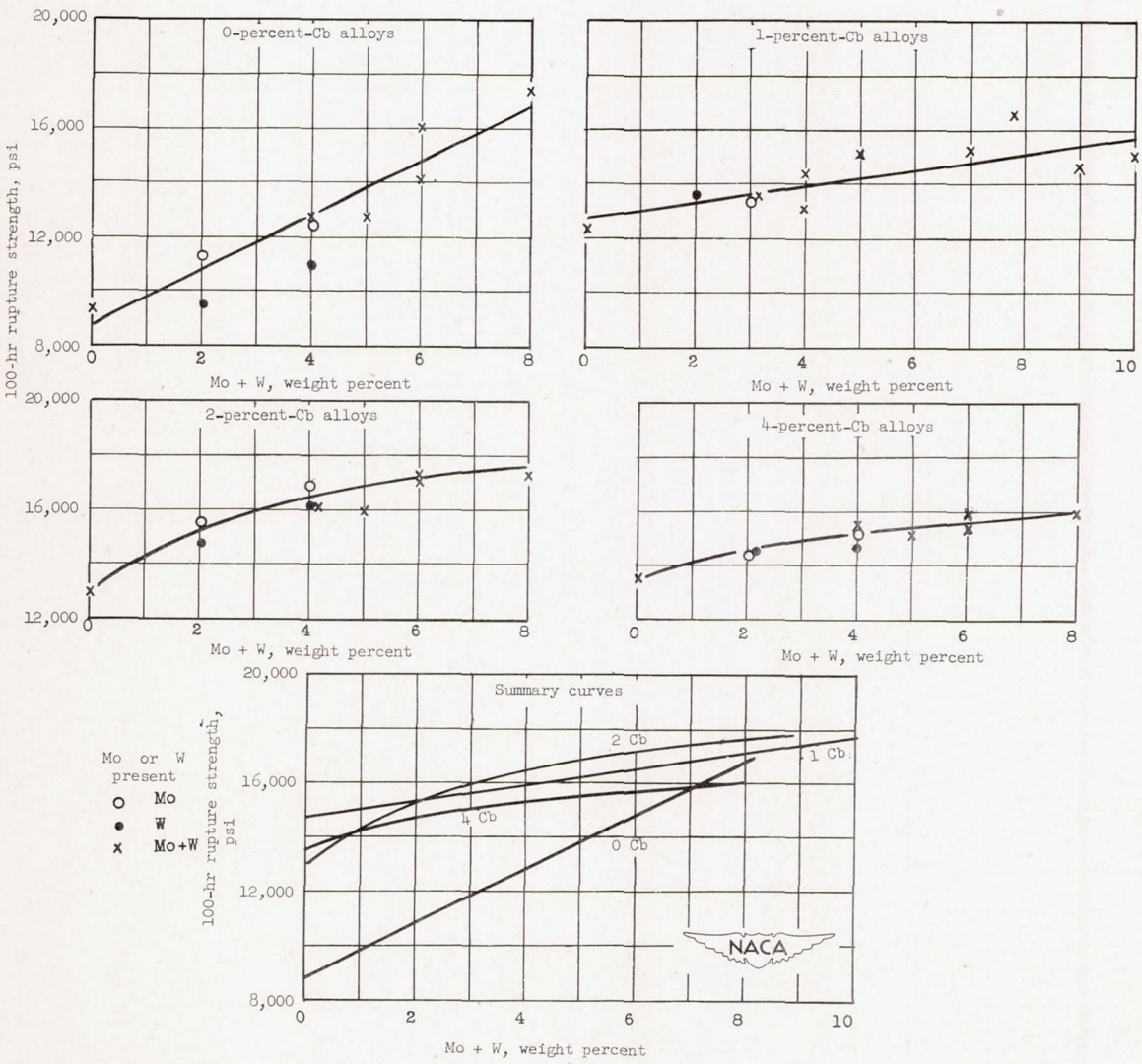


Figure 23.- Influence of molybdenum and tungsten additions on a total weight percent basis on 100-hour rupture strength at 1500° F of columbium modifications.

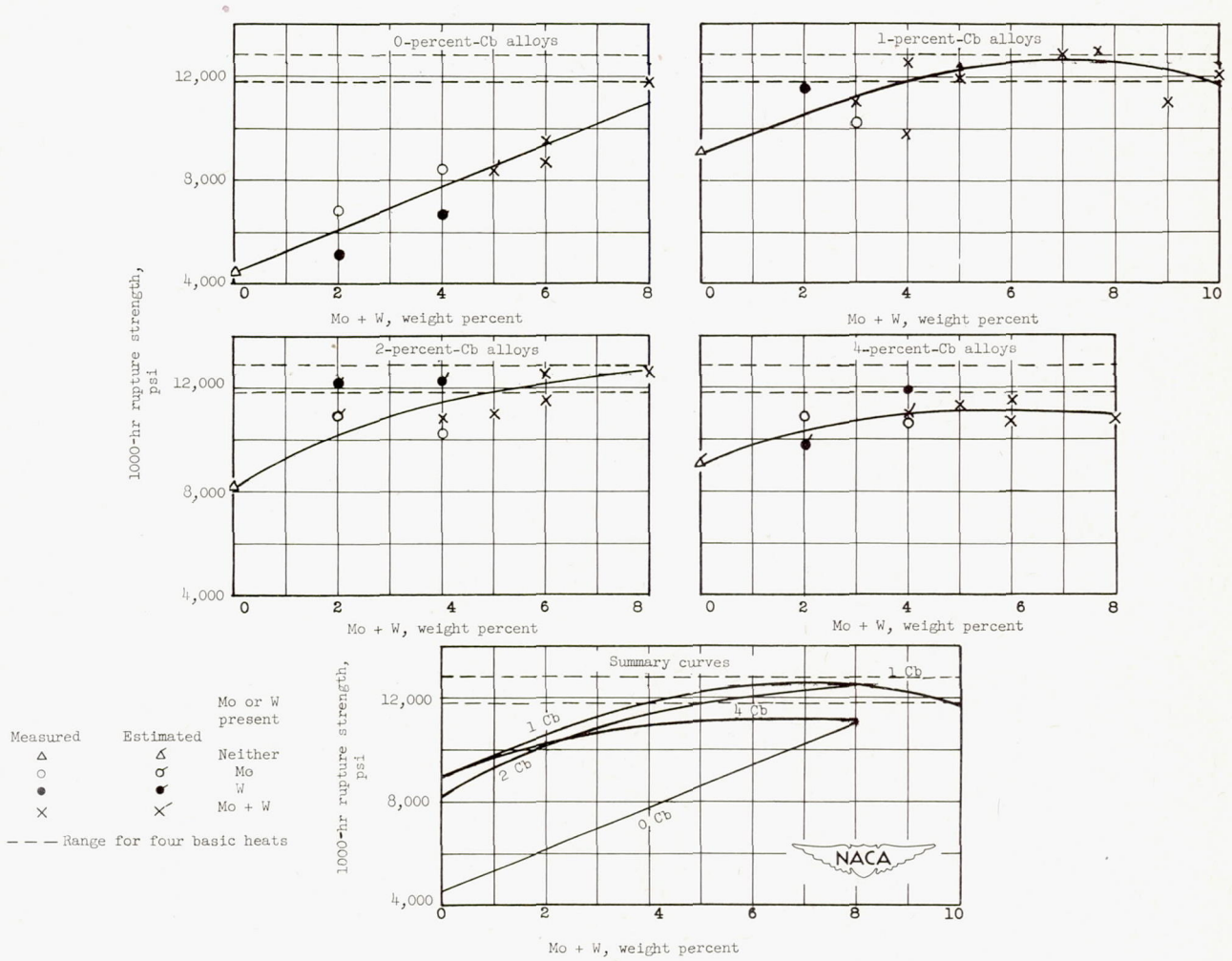
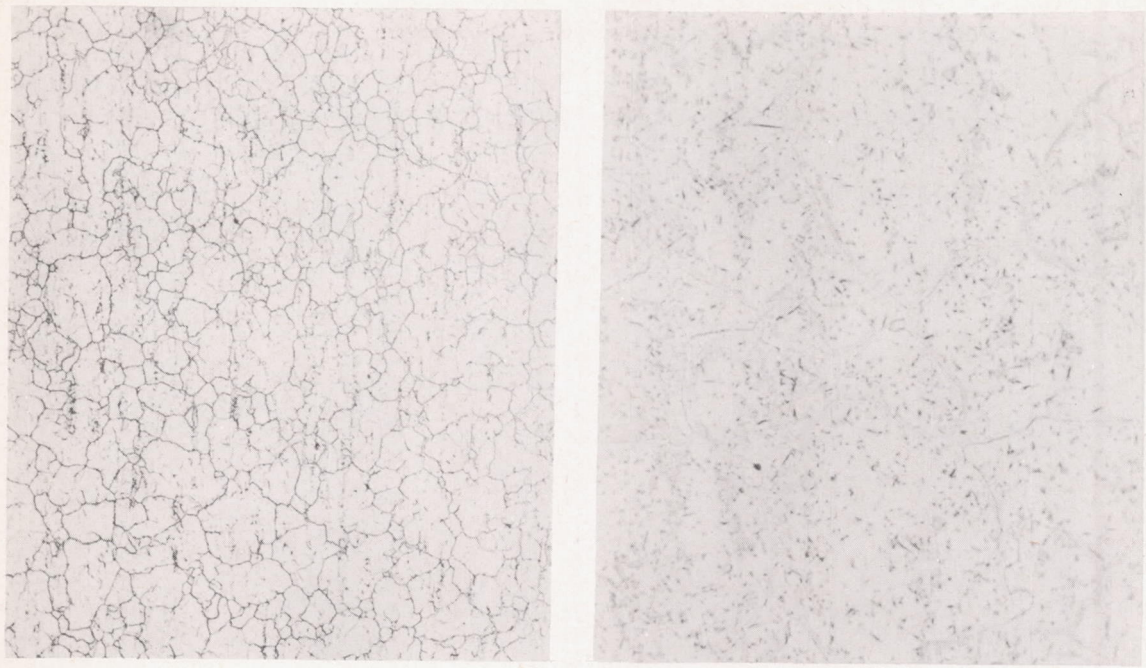
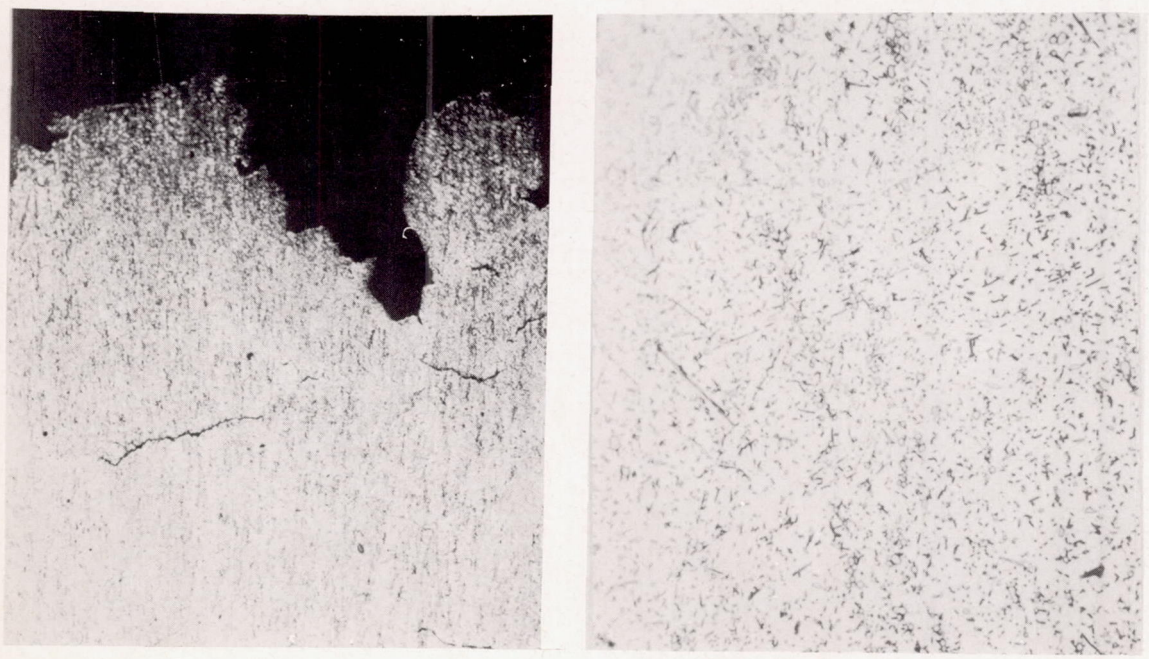


Figure 24.- Influence of molybdenum and tungsten additions on a total weight percent basis on 1000-hour rupture strength at 1500° F of columbium modifications.



100X 1000X

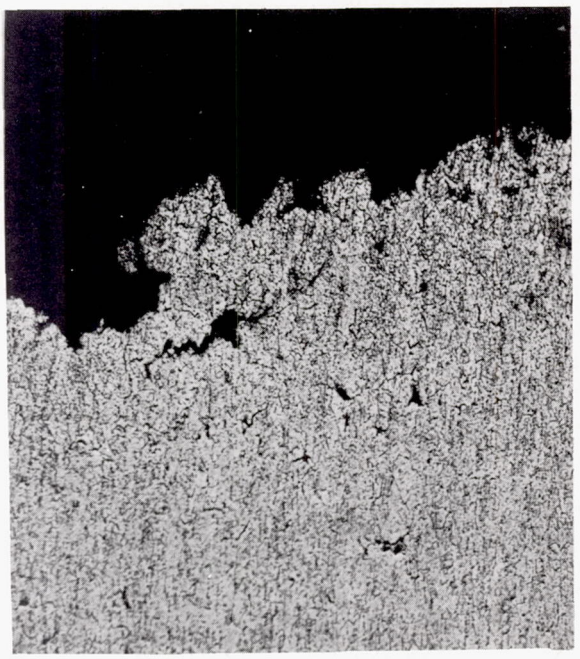
(a) Original: Solution-treated at 2200° F for 1 hour and water-quenched; aged at 1400° F for 24 hours and air-cooled.



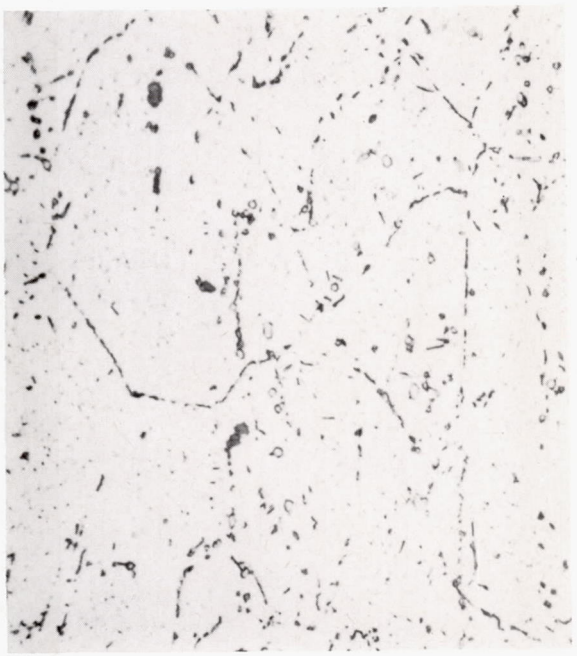
100X 1000X

(b) Rupture specimen: 446 hours for rupture under 12,000 psi at 1500° F.

Figure 25.- Microstructures of basic alloy (heat 8).

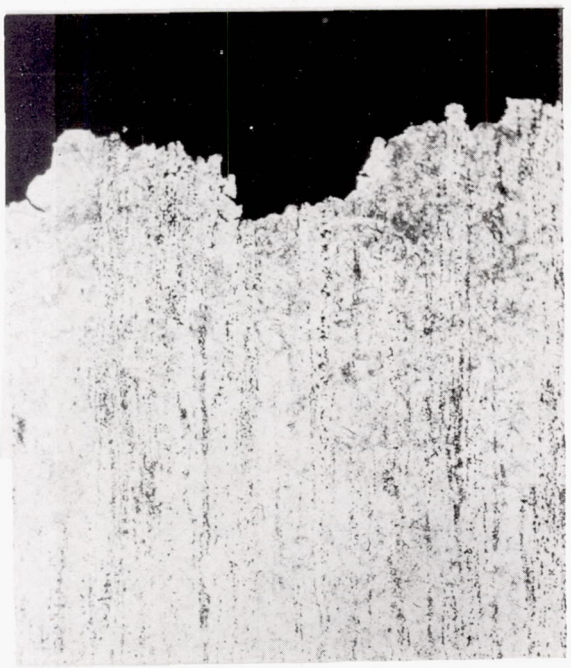


100X



1000X

(a) Carbon, 0.08 percent (heat 13); 328 hours for rupture under 13,000 psi.



100X



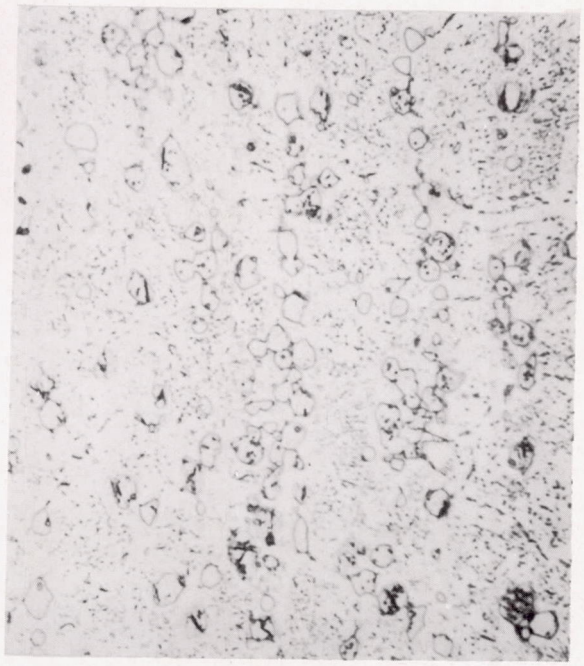
1000X

(b) Carbon, 0.40 percent (heat 15); 407 hours for rupture under 15,000 psi.

Figure 26.- Microstructures of carbon modifications of basic alloy after rupture testing at 1500° F.



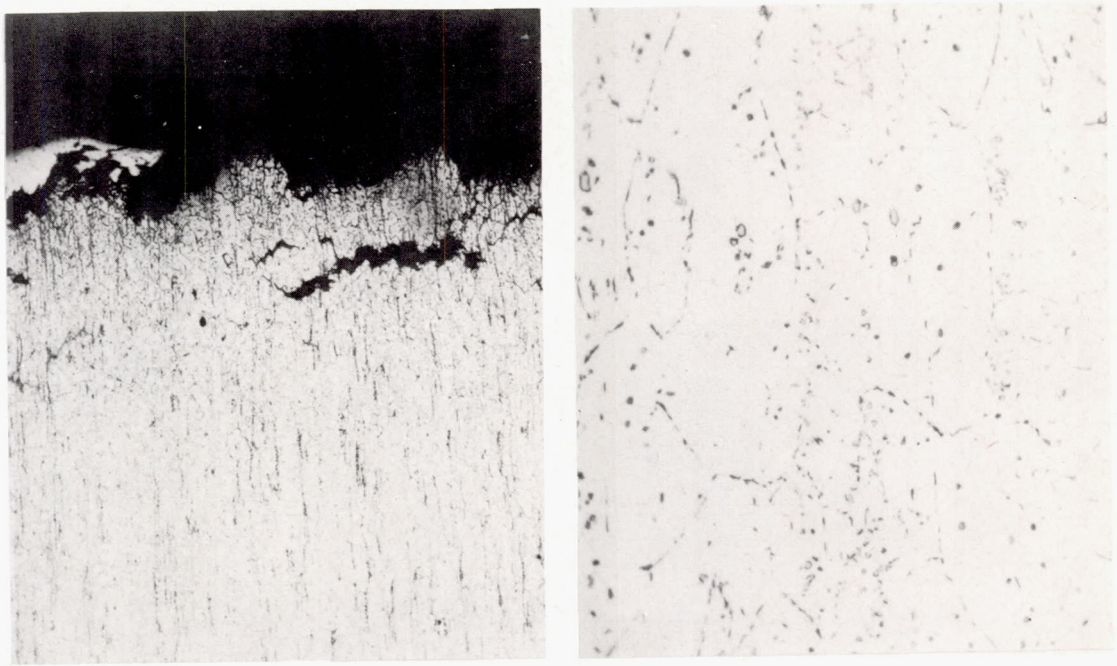
100X



1000X

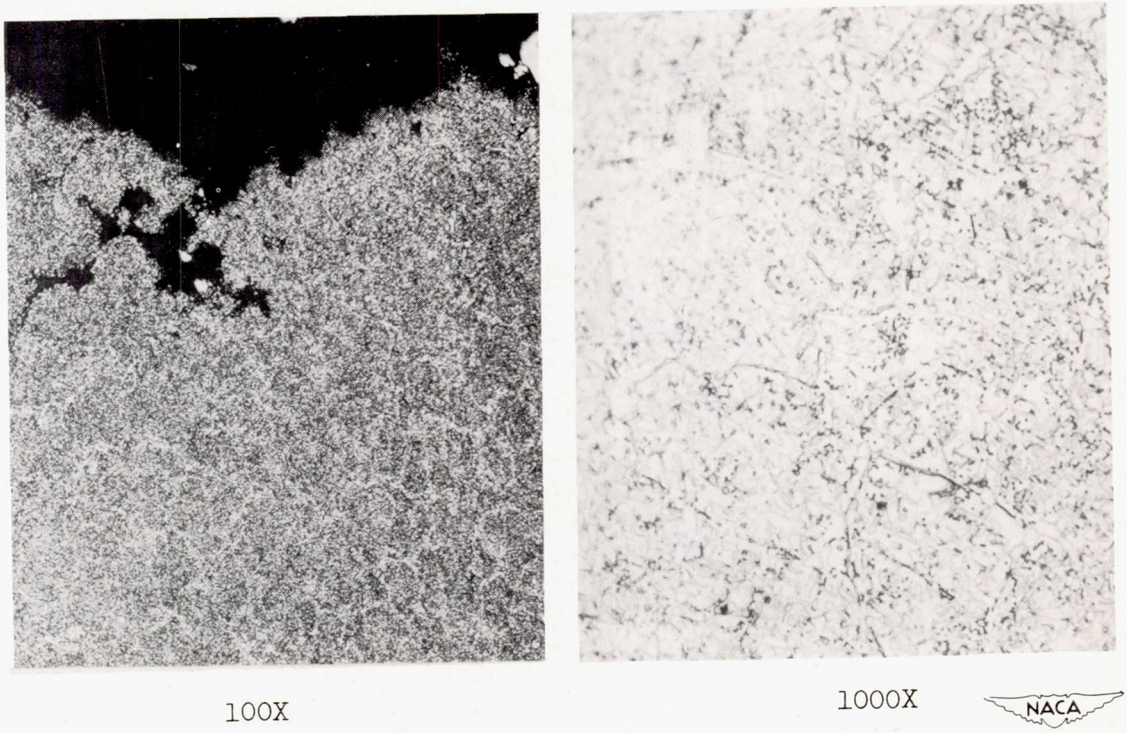
(c) Carbon, 0.60 percent (heat 16); 494 hours for rupture under 12,000 psi.

Figure 26.- Concluded.



100X 1000X

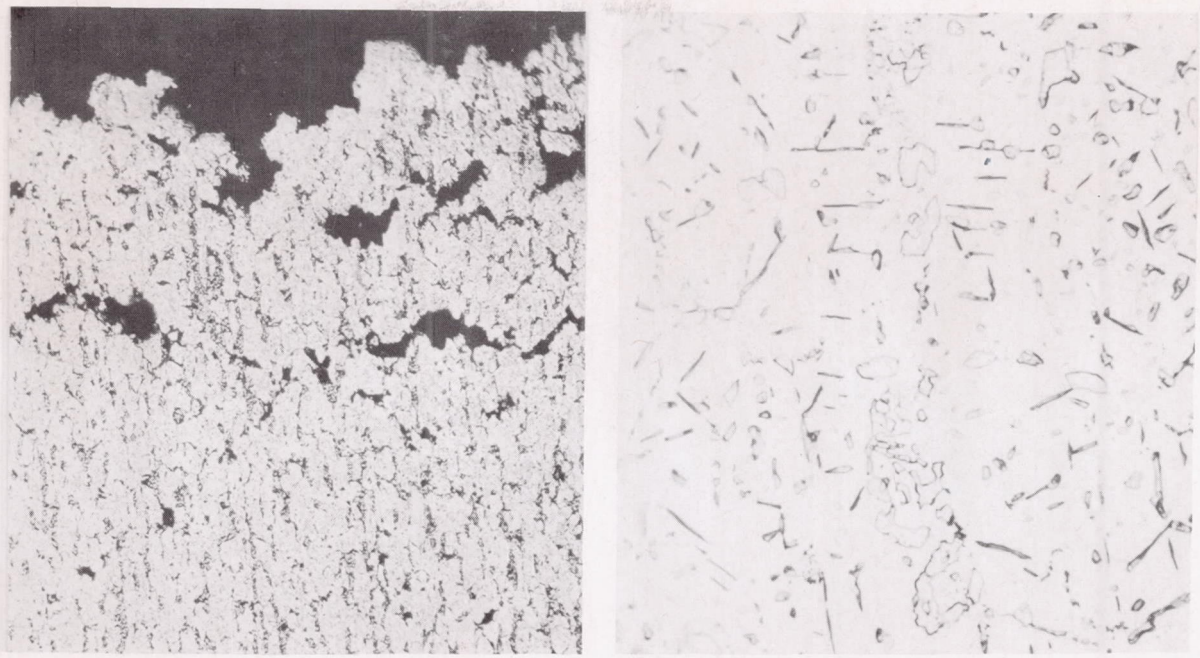
(a) Chromium, 10 percent (heat 51); 318 hours for rupture under 12,000 psi.



100X 1000X

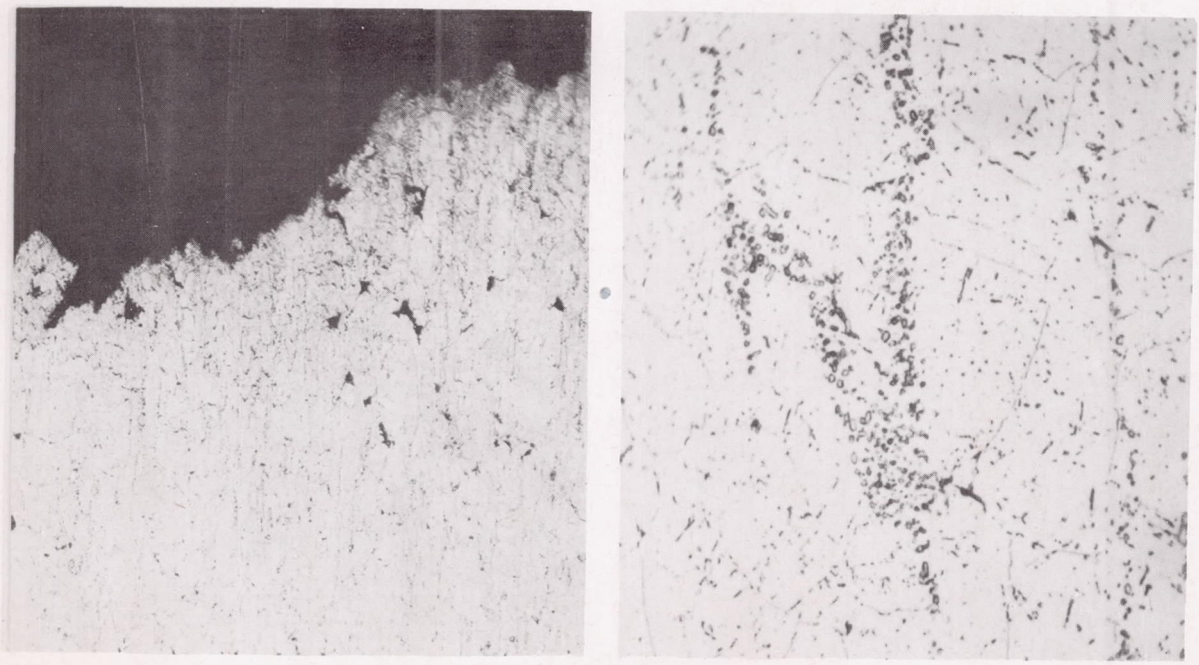
(b) Chromium, 30 percent (heat 52); 363 hours for rupture under 14,000 psi.

Figure 27.- Microstructures of chromium modifications of basic alloy after rupture testing at 1500° F.



100X 1000X

(a) Nickel, 0 percent (heat 23); 286 hours for rupture under 10,500 psi.



100X 1000X

(b) Nickel, 10 percent (heat 25); 516 hours for rupture under 15,500 psi.

Figure 28.- Microstructures of nickel modifications of basic alloy after rupture testing at 1500° F.



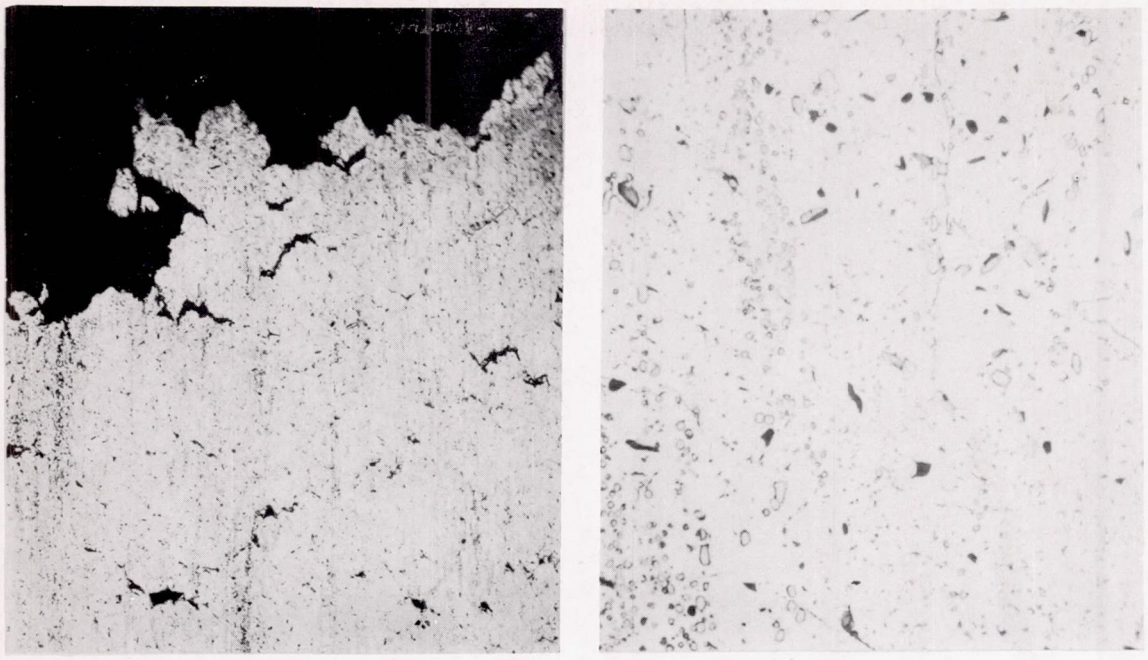
100X



1000X

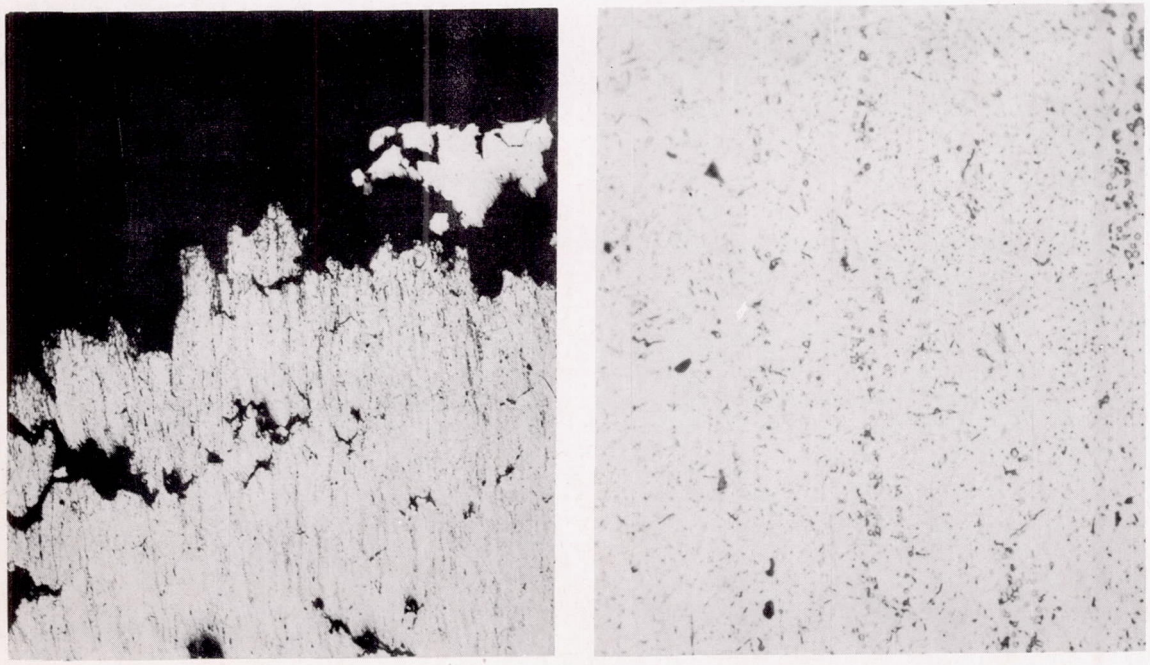
(c) Nickel, 30 percent (heat 26); 594 hours for rupture under 12,500 psi.

Figure 28.- Concluded.



100X 1000X

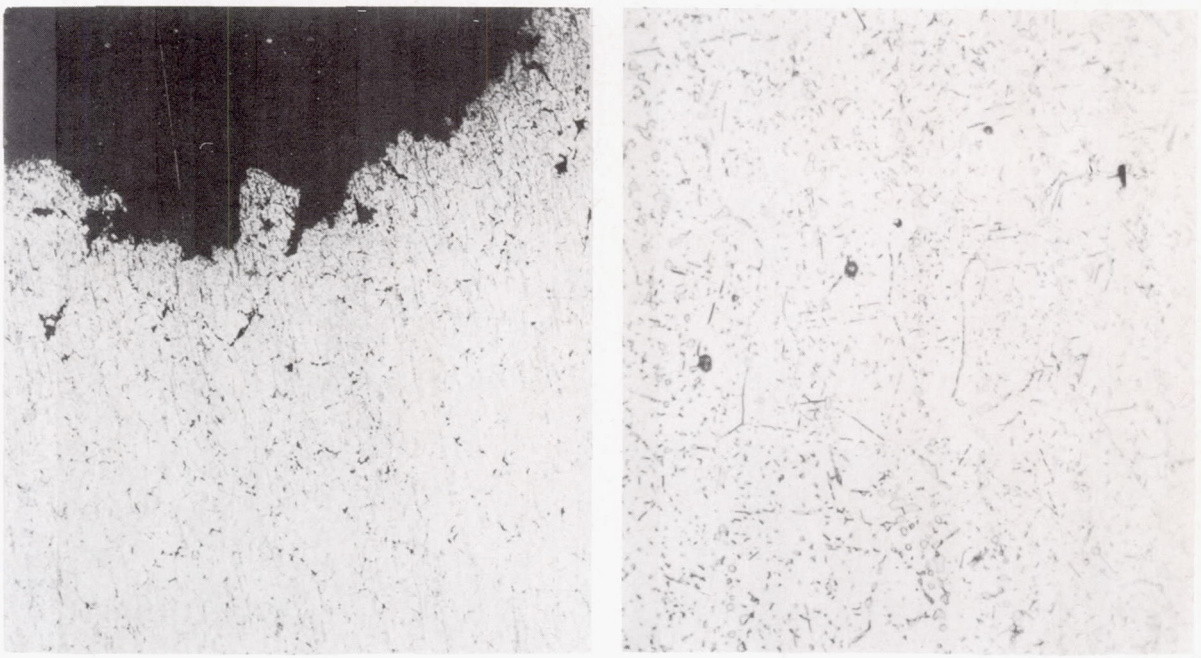
(a) Cobalt, 0 percent (heat 29); 330 hours for rupture under 12,500 psi.



100X 1000X

(b) Cobalt, 10 percent (heat 30); 446 hours for rupture under 13,500 psi.

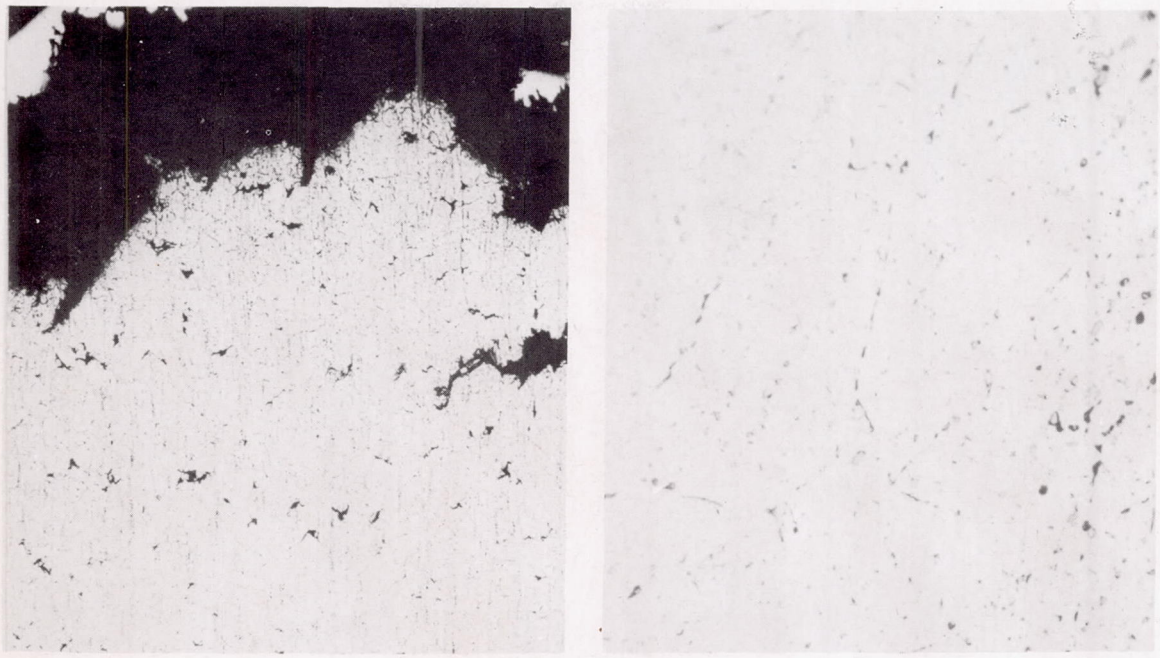
Figure 29.- Microstructures of cobalt modifications of basic alloy after rupture testing at 1500° F.



100X 1000X

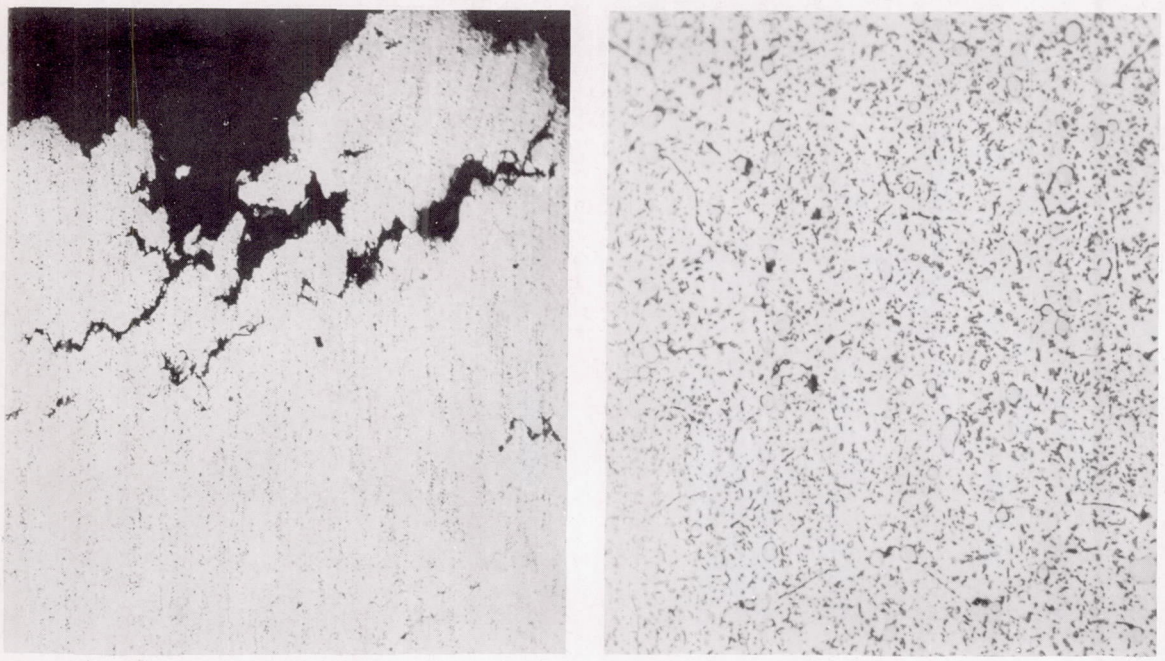
(c) Cobalt, 32 percent (heat 31); 409 hours for rupture under 13,000 psi.

Figure 29.- Concluded.



100X 1000X

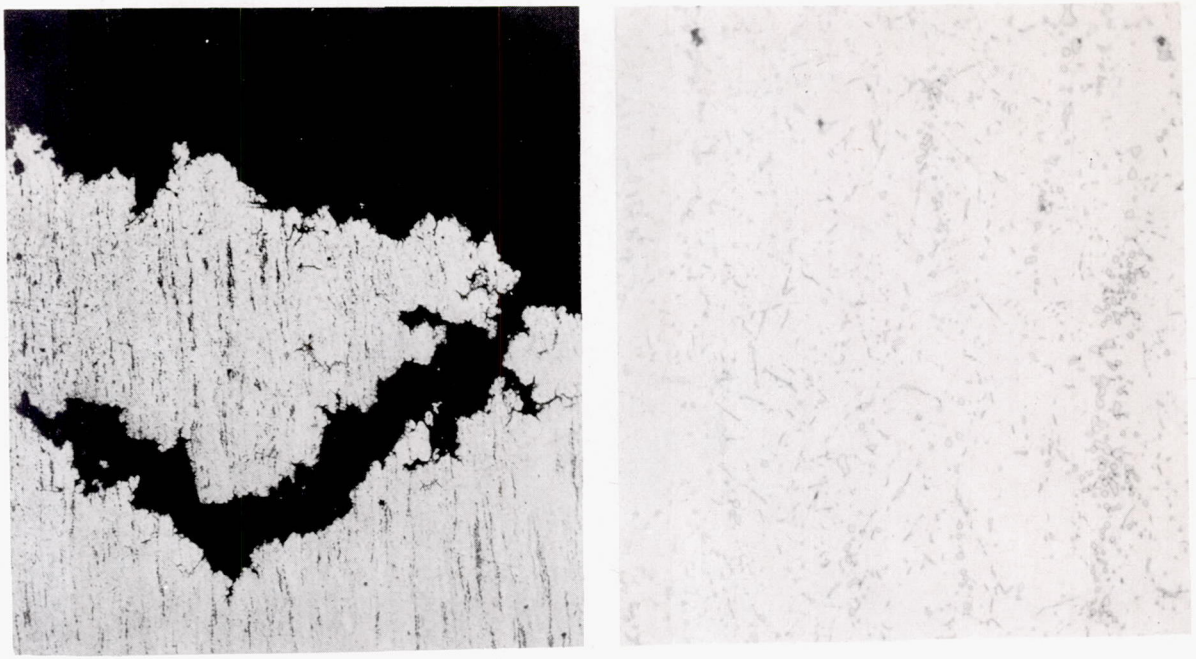
(a) Molybdenum, 0 percent (heat 32); 539 hours for rupture under 12,500 psi.



100X 1000X

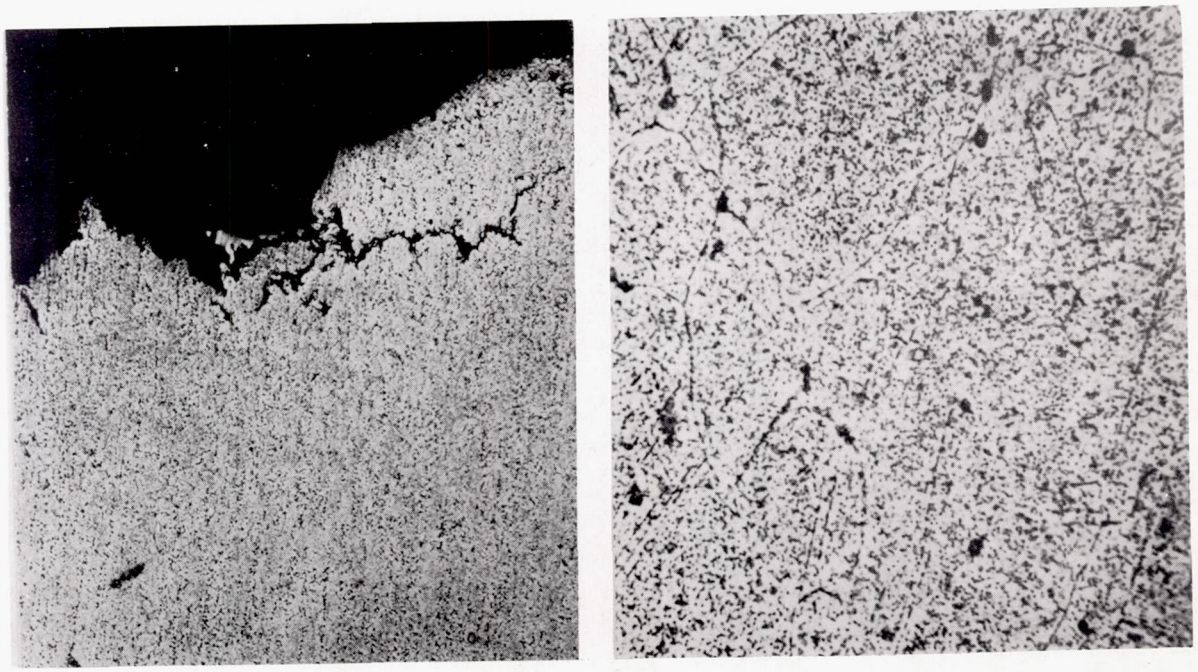
(b) Molybdenum, 7 percent (heat 36); 397 hours for rupture under 13,000 psi.

Figure 30.- Microstructures of molybdenum modifications of basic alloy after rupture testing at 1500° F.



100X 1000X

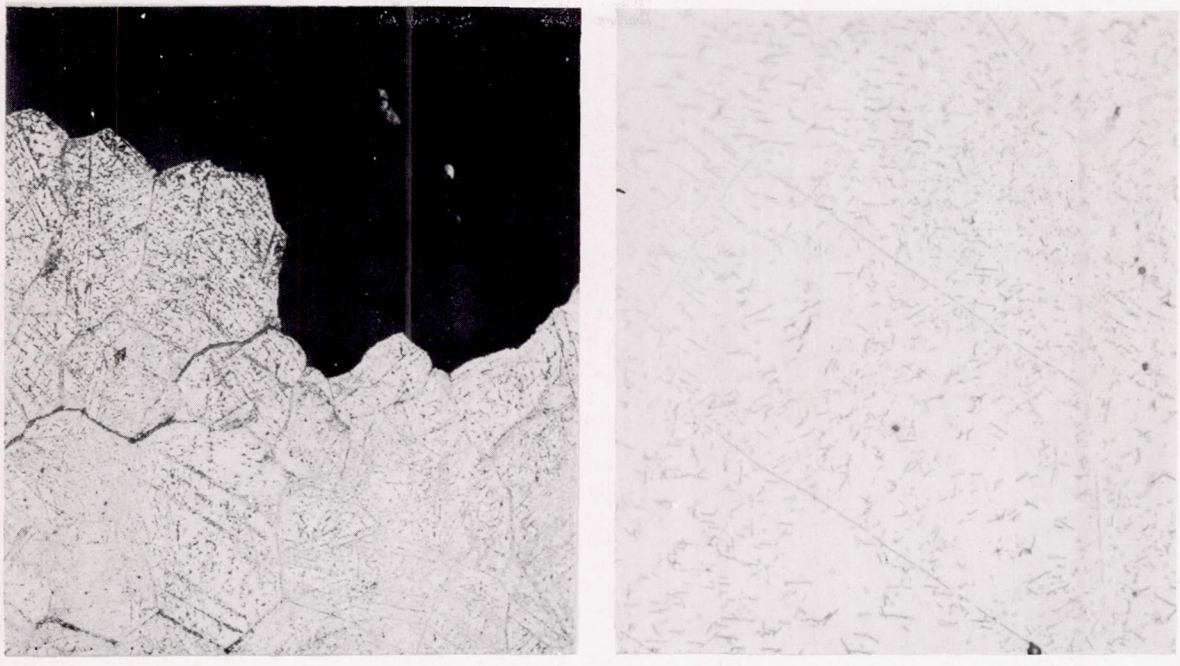
(a) Tungsten, 0 percent (heat 37); 292 hours for rupture under 13,500 psi.



100X 1000X

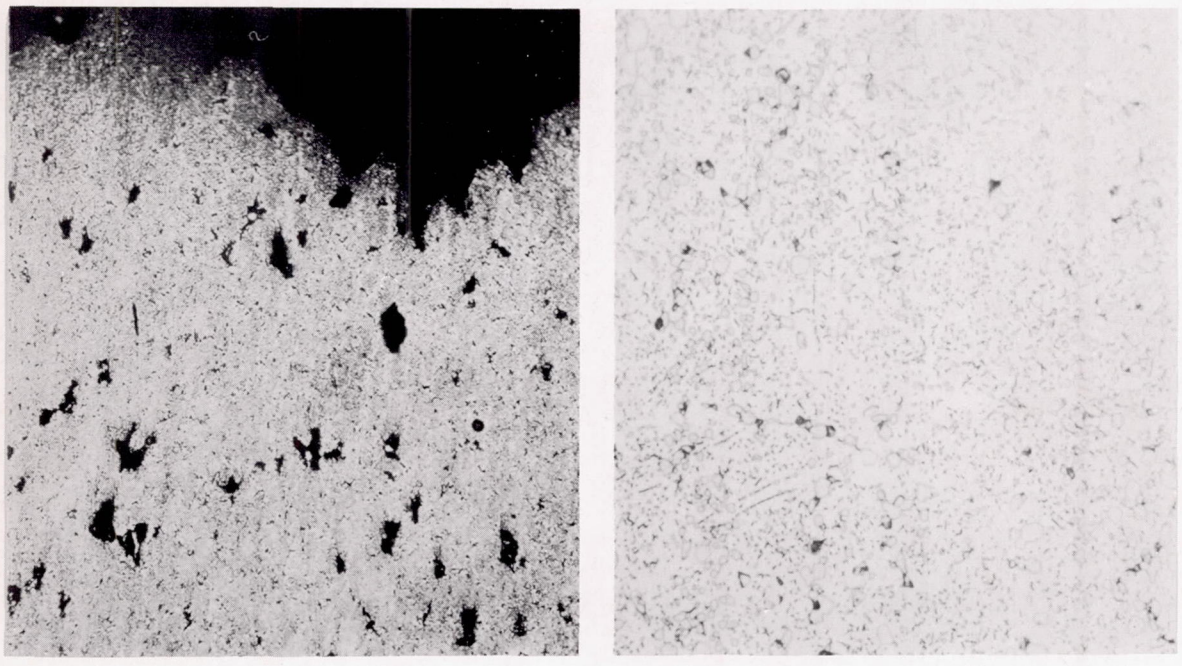
(b) Tungsten, 7 percent (heat 40); 209 hours for rupture under 16,500 psi.

Figure 31.- Microstructures of tungsten modifications of basic alloy after rupture testing at 1500° F.



100X 1000X

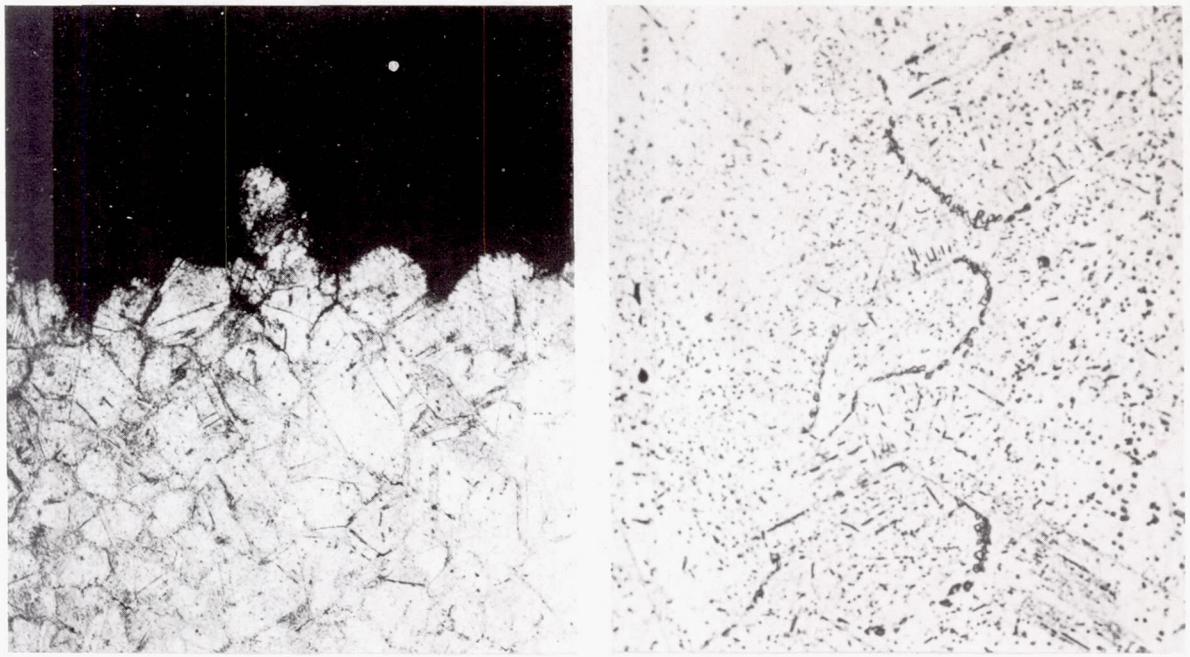
(a) Columbium, 0 percent (heat 47); 389 hours for rupture under 10,000 psi.



100X 1000X

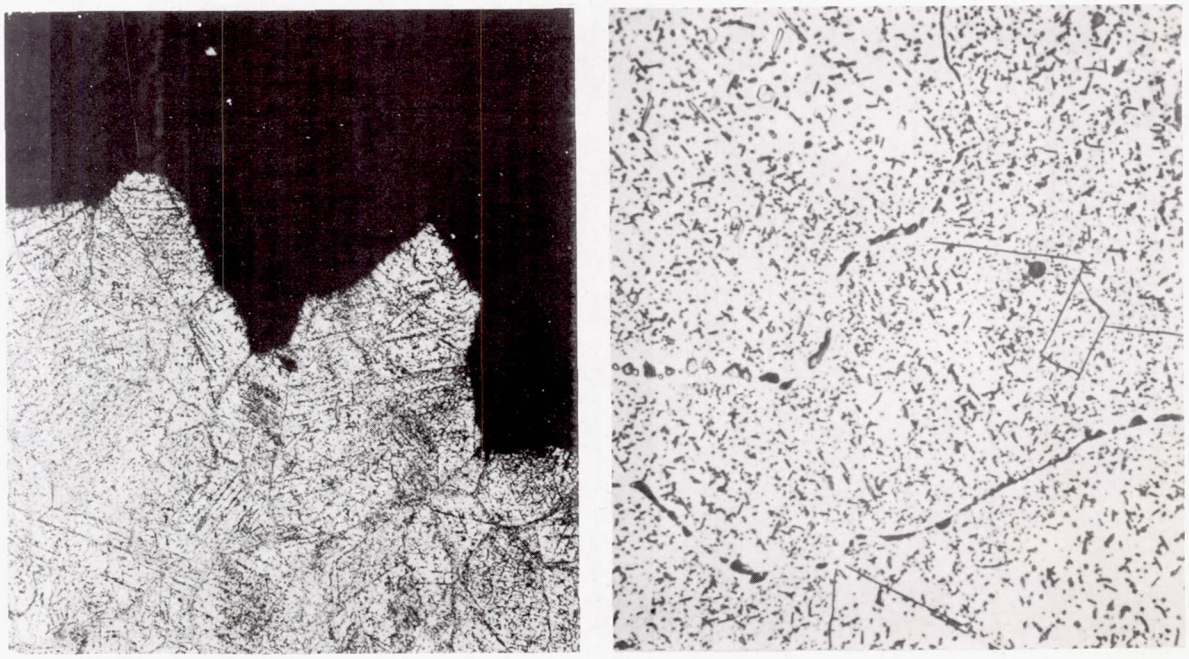
(b) Columbium, 6 percent (heat 50); 592 hours for rupture under 12,000 psi.

Figure 32.- Microstructures of columbium modifications of basic alloy after rupture testing at 1500° F.



100X 1000X

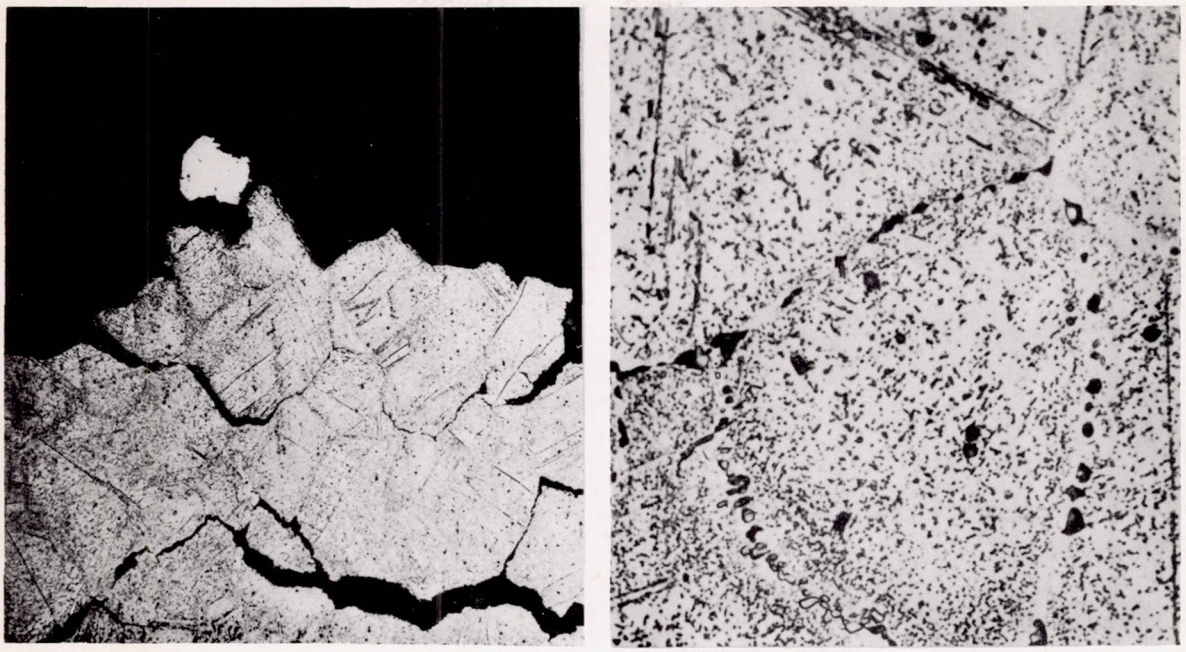
(a) 2Mo-0W-0Cb (heat 45); 385 hours for rupture under 8500 psi.



100X 1000X

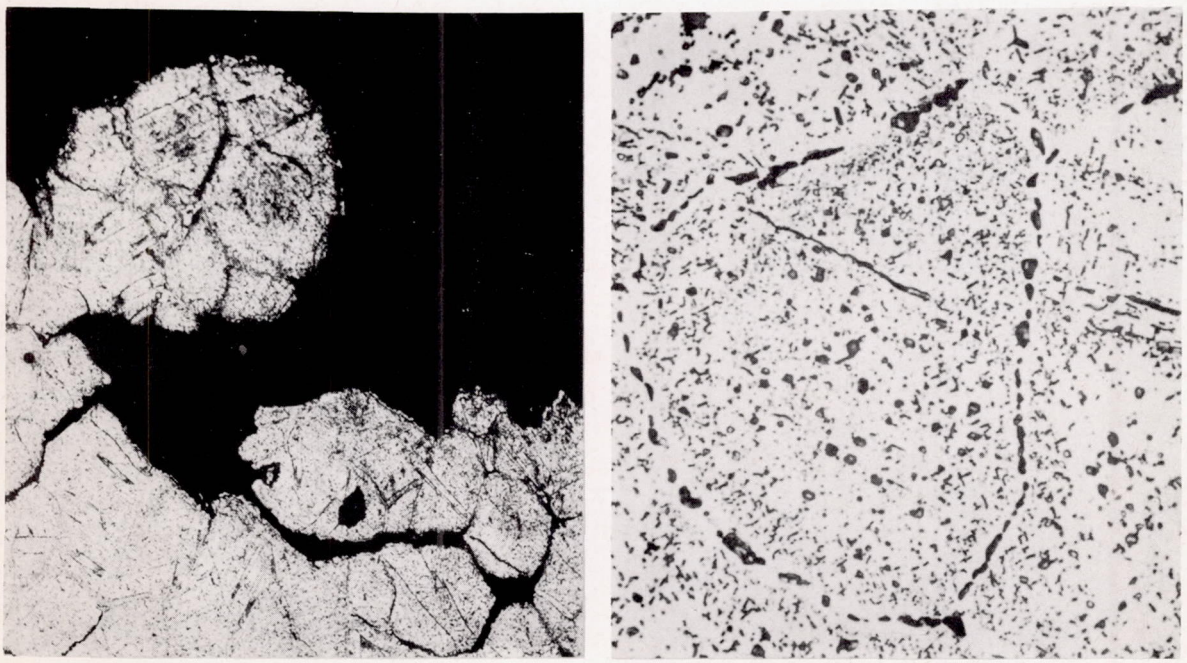
(b) 4Mo-0W-0Cb (heat 46); 334 hours for rupture under 10,000 psi.

Figure 33.- Microstructures of molybdenum or tungsten alloys after rupture testing at 1500° F.



100X 1000X

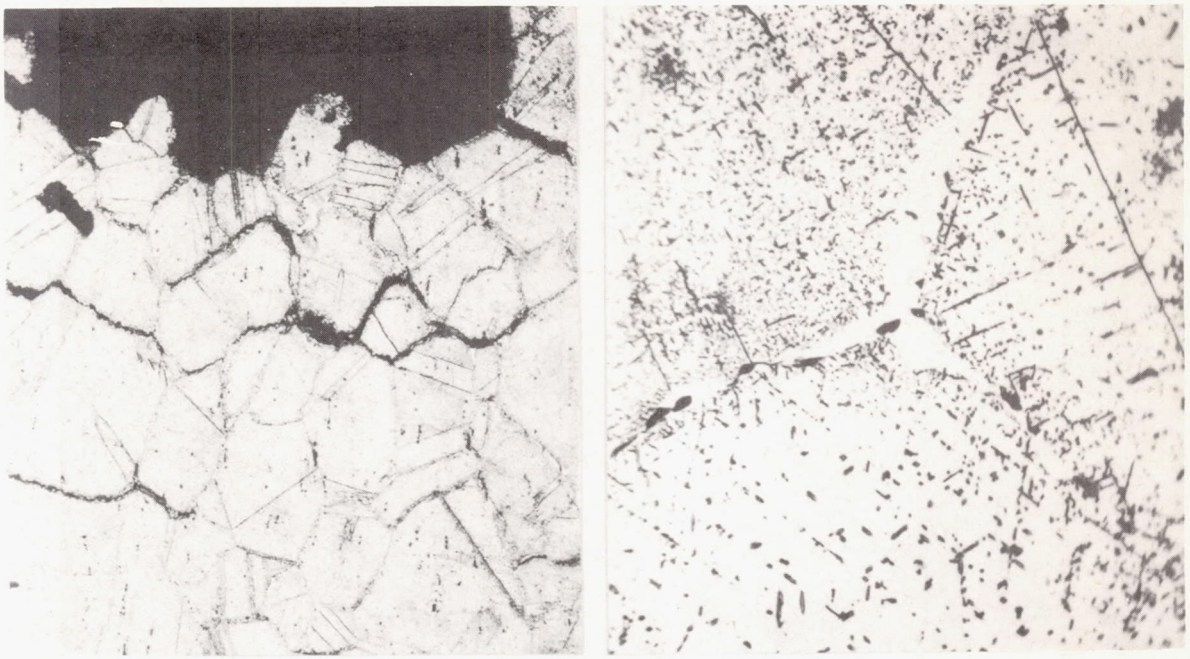
(c) 0Mo-2W-0Cb (heat 53); 547 hours for rupture under 6000 psi.



100X 1000X

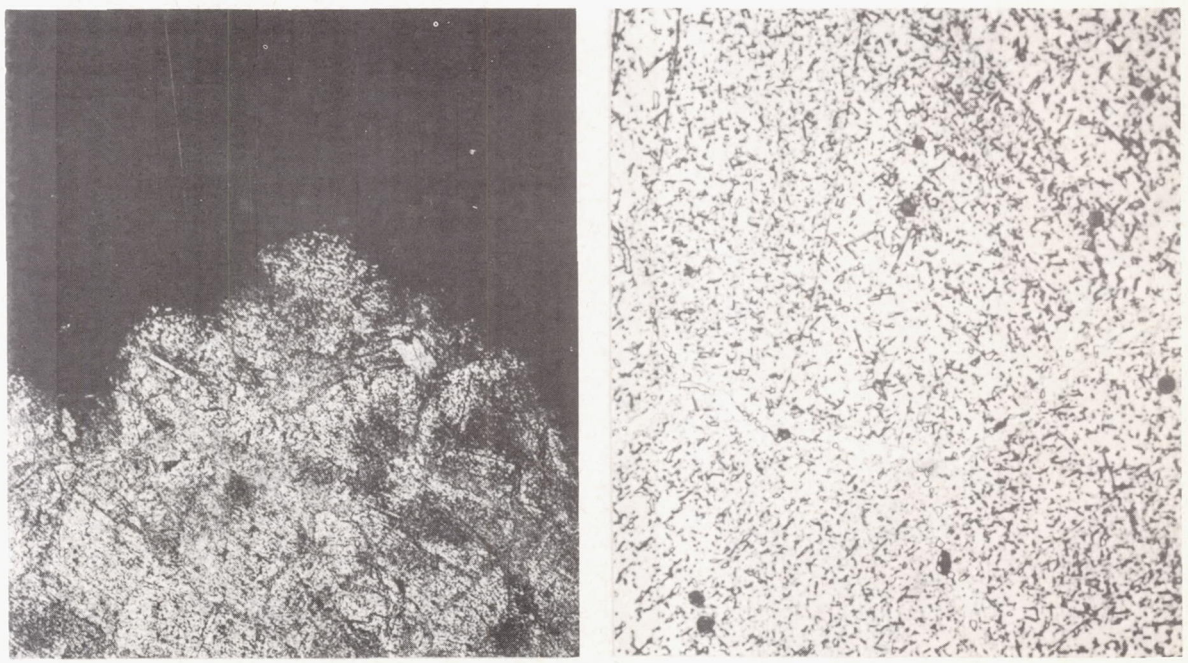
(d) 0Mo-4W-0Cb (heat 54); 1179 hours for rupture under 6500 psi.

Figure 33.- Concluded.



100X 1000X

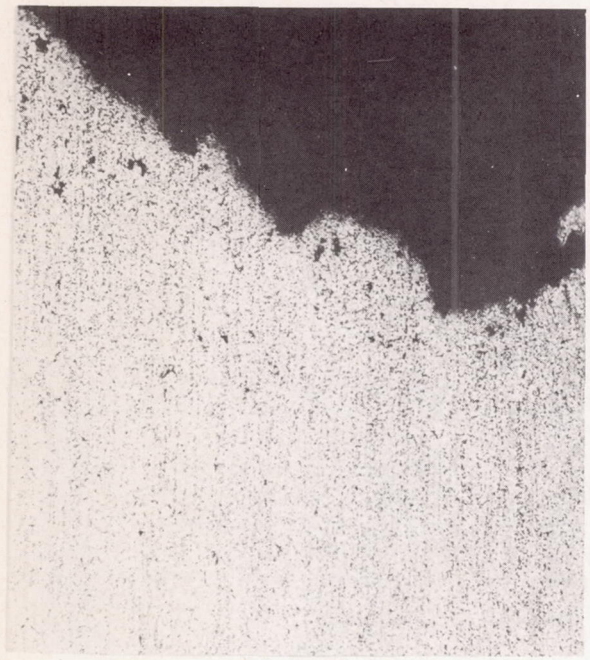
(a) 0Mo-0W-0Cb (heat 43); 390 hours for rupture under 6000 psi.



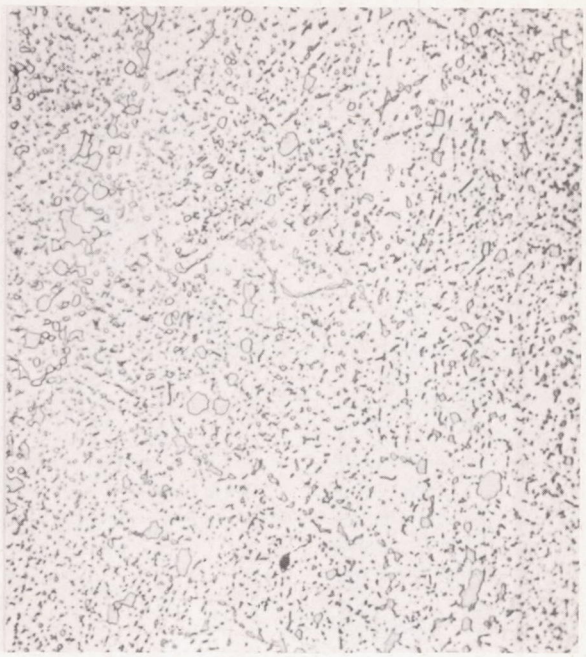
100X 1000X

(b) 4Mo-4W-0Cb (heat 58); 551 hours for rupture under 13,000 psi.

Figure 34.- Influence of molybdenum plus tungsten plus columbium content on microstructure after rupture testing at 1500° F.

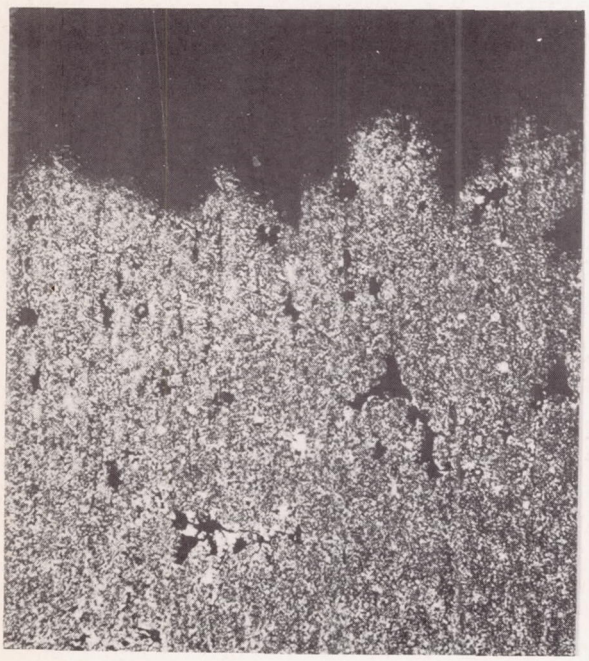


100X



1000X

(c) 4Mo-4W-2Cb (heat 84); 579 hours for rupture under 13,500 psi.



100X



1000X

(d) 4Mo-4W-4Cb (heat 67); 545 hours for rupture under 12,000 psi.

Figure 34.- Concluded.

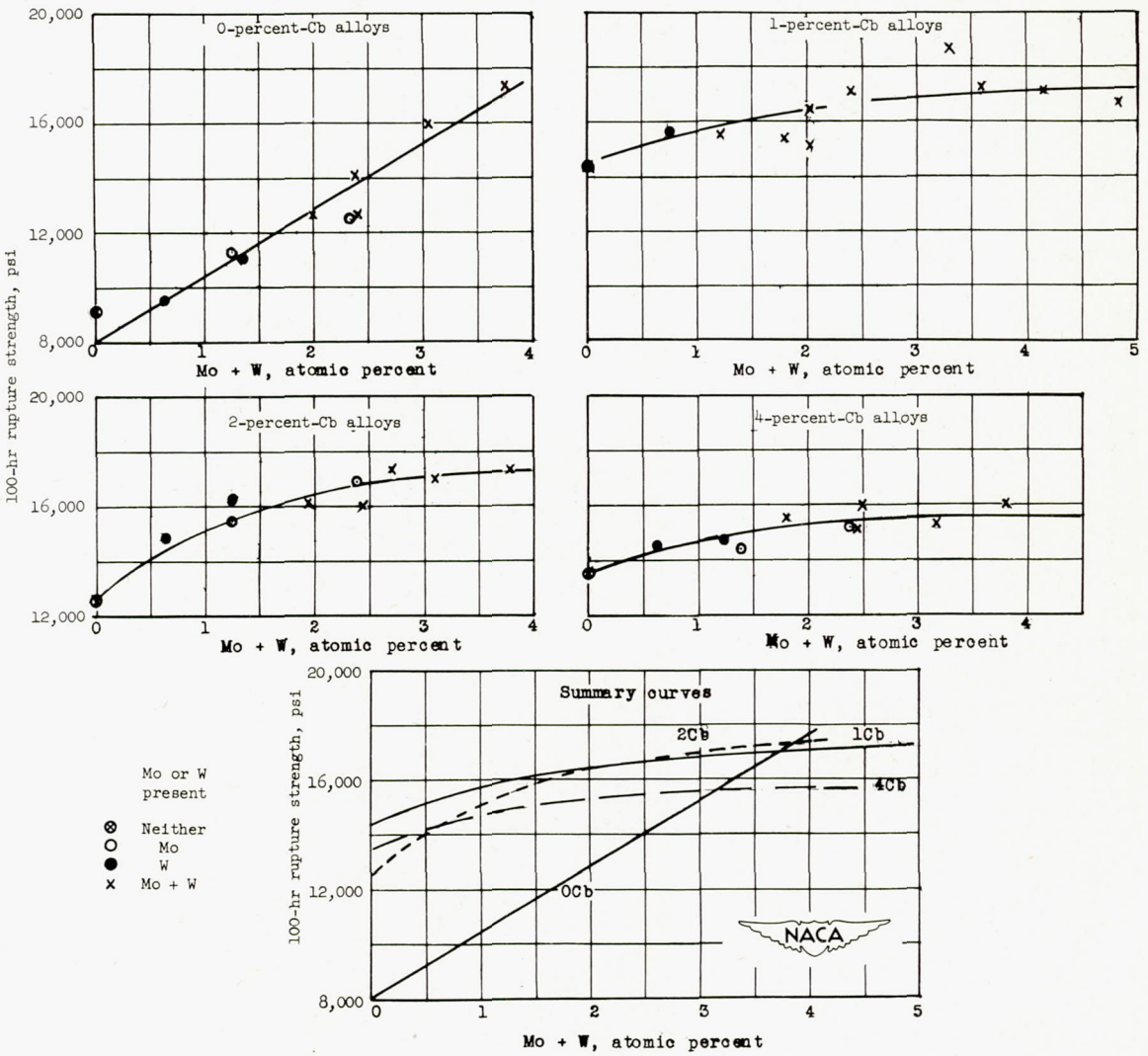


Figure 35.- Influence of molybdenum and tungsten additions on a total atomic percent basis on 100-hour rupture strength at 1500° F of columbium modifications.

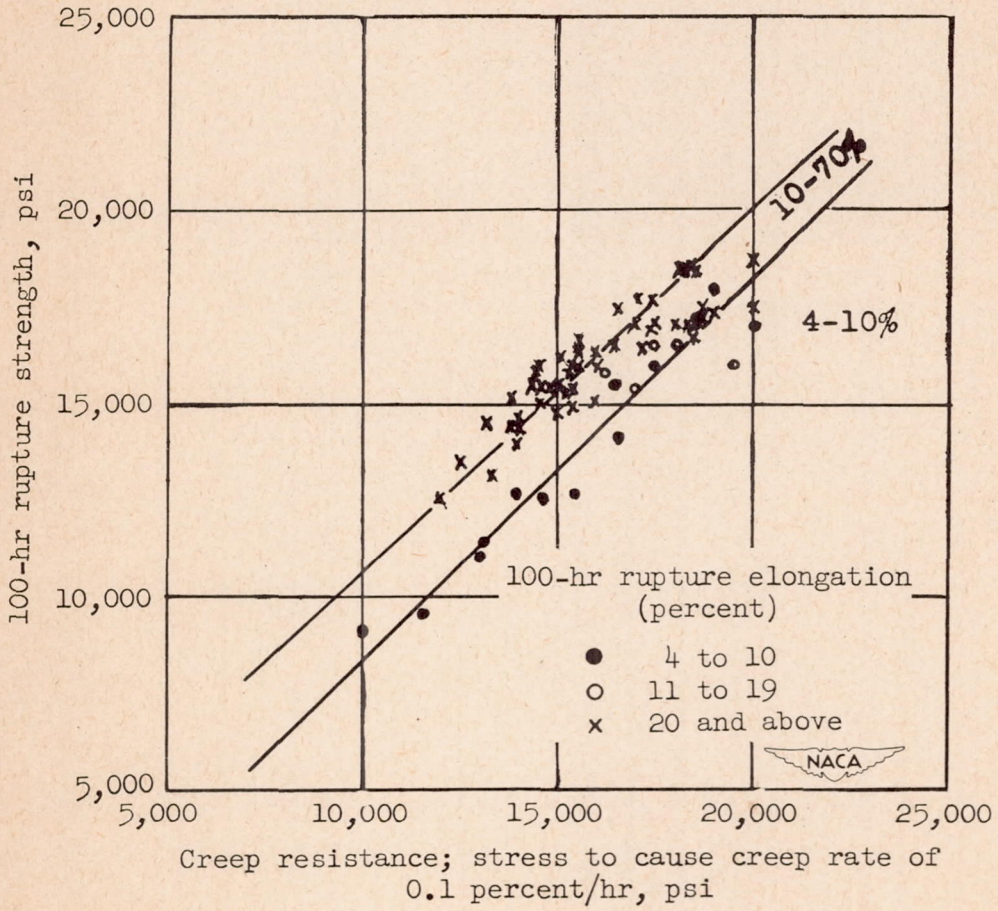


Figure 36.- Influence of creep resistance and 100-hour rupture elongation on the 100-hour rupture strength at 1500° F of modified alloys.